JOHN G. McMULLIN

JOURNAL OF Metals

TECHNOLOGY · PRACTICE

MAY 1949

Buyers' Market

•

ace Metals in Farming

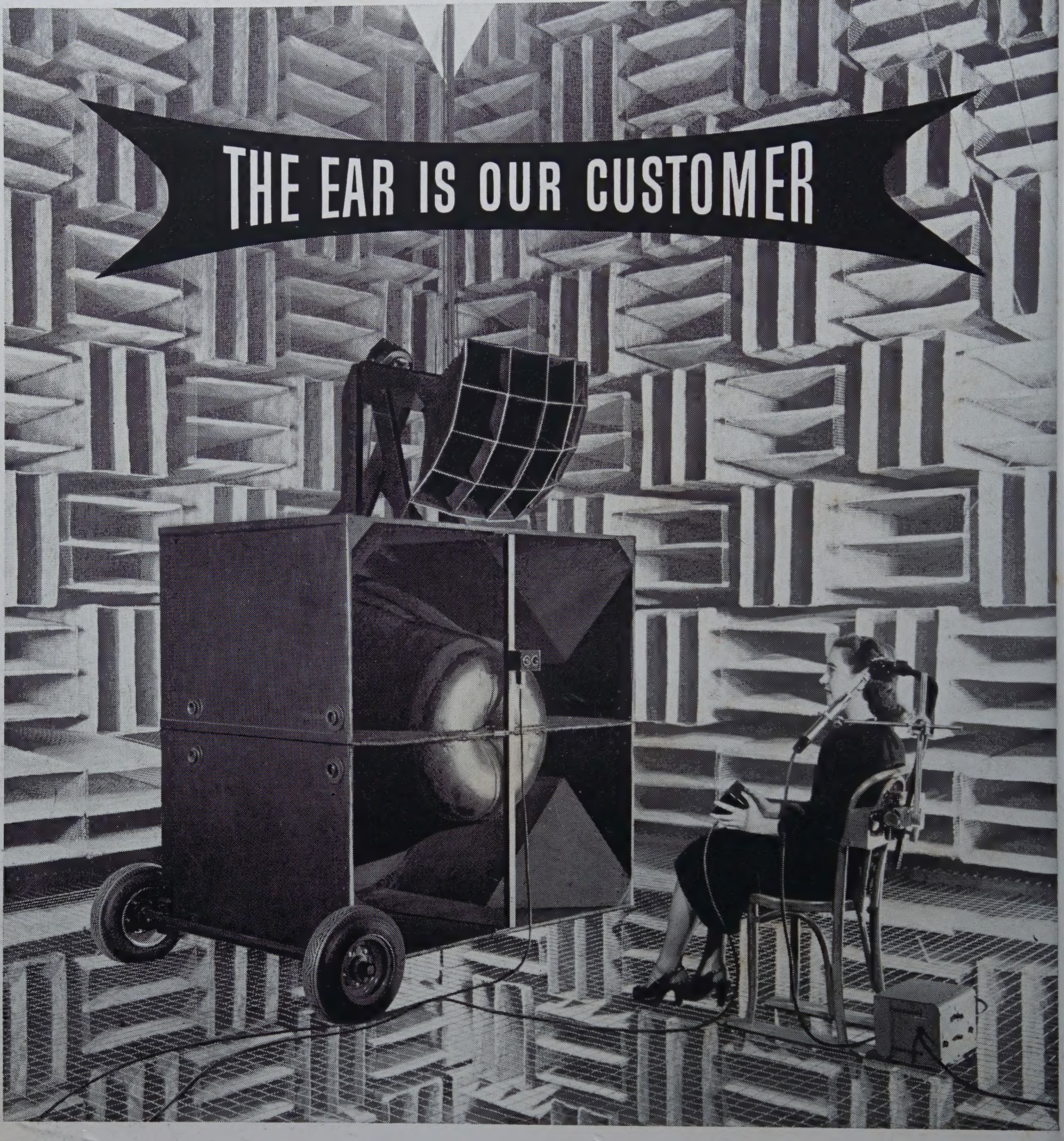
•

st Furnace Conference

•

D Discussion—Fall, '48

A MONTHLY PUBLICATION OF THE
AMERICAN INSTITUTE OF MINING AND METALLURGICAL ENGINEERS
METALS BRANCH



What happens when you hear? What happens *inside* your ear when sound waves come in from a telephone conversation?

Bell Telephone Laboratories scientists have developed special apparatus to help answer these questions, for the telephone system is designed to meet the ear's requirements for good listening.

In the test pictured above, the young lady sits before loudspeakers in a soundproofed room with a small hollow tube, reaching just inside the ear canal. Sounds differing slightly in frequency and intensity come from a loudspeaker. The subject seeks to tell one from another, recording her judgment electrically by pressing a switch.

Meanwhile, the same sound waves pass down the hollow tube to a condenser microphone, and a record is made of the exact sound intensities she identified. Results help reveal the sound levels you can hear clearly and without strain—the sounds your telephone must be designed to carry.

Scientists at Bell Telephone Laboratories make hundreds of tests in this manner. It's just one part of the work which goes on year after year at the Laboratories to help keep Bell System telephone service the finest on earth.

BELL TELEPHONE LABORATORIES

Exploring and inventing, devising and perfecting, for continued improvements and economies in telephone service.

Control the Air and you control the Furnace

Twice as many tons of air pass through the furnace as tons of steel

This is why **air** is the controlling factor in furnace operation. The ISLEY system supplies and controls the air. The supply is positive. The volume supplied by the fan goes to the burners. No leaky reversing valves and dampers interfere with the accuracy of controls.

The ISLEY draft control is mechanical, positive and independent of the temperature of regenerators, stacks and weather. There are no leaks to dilute the waste gas and cool it.

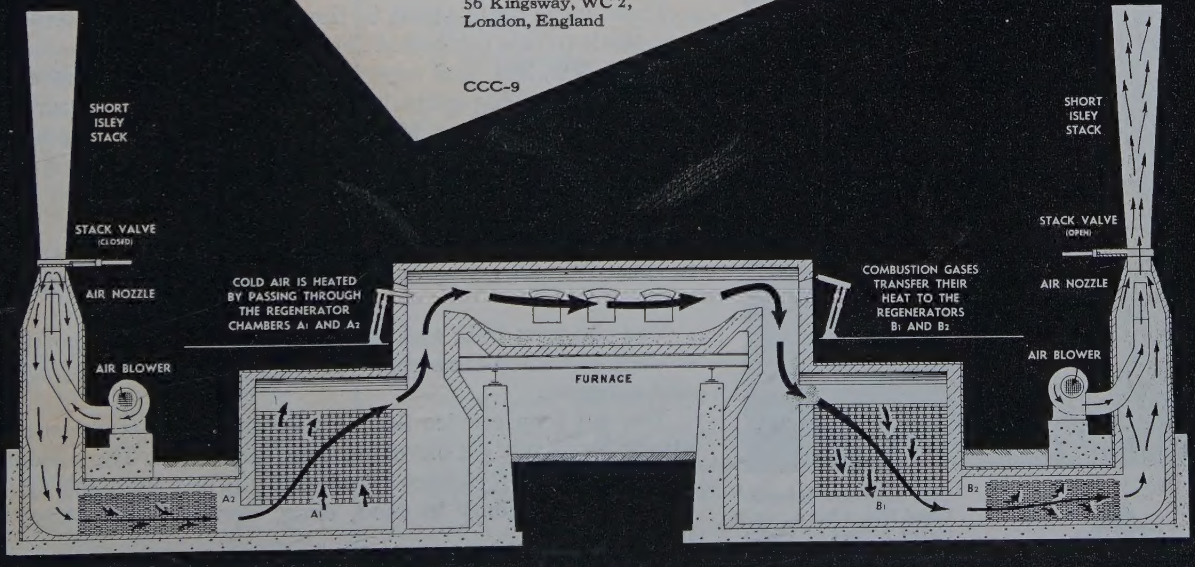
For more uniform furnace pressure, perfect control of combustion air, better heat recovery from waste gas, let us give you the complete ISLEY story.

MORGAN
CONSTRUCTION CO.
WORCESTER, MASS.

Pittsburgh Office:
2815 Koppers Building

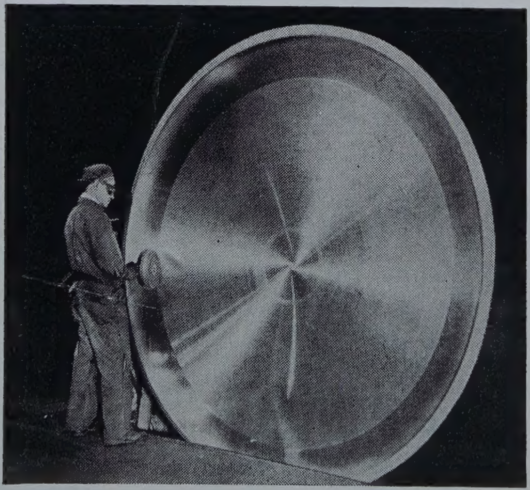
English Representative:
International Construction Co.,
56 Kingsway, WC 2,
London, England

CCC-9


SHORT ISLEY STACK
STACK VALVE (CLOSED)
AIR NOZZLE
AIR BLOWER
COLD AIR IS HEATED BY PASSING THROUGH THE REGENERATOR CHAMBERS A1 AND A2
A2
A1
FURNACE
COMBUSTION GASES TRANSFER THEIR HEAT TO THE REGENERATORS B1 AND B2
B1
B2
SHORT ISLEY STACK
STACK VALVE (OPEN)
AIR NOZZLE
AIR BLOWER

ISLEY FURNACE CONTROL SYSTEM

This large stainless steel tank end being spun to the desired radius is made up of standard width sheets of "18-8" welded together and the welds ground smooth.
(Below) Another large stainless steel tank end, already spun to shape, shown here being polished.



HERE is austenitic chromium-nickel stainless steel, Type 305, being spun into a huge tank end. Another, of Type 302 . . . spun to shape and receiving a polish . . . is pictured at the left.

These large units attest to the adaptability of austenitic stainless steels to this method of fabrication.

Types 302, 304, and especially Type 305, lend themselves well to modern spinning operations according to the producers of these tank ends, Milwaukee Metal Spinning Company (Spincraft) of Milwaukee 8, Wisc. On a production basis, this company spins heads and ends for milk storage tanks ranging from 36 to 102 inches in diameter.

This is only one of many applications in which advantage may be taken of the various useful properties of chromium-nickel stainless steels.

You can trim bulk and deadweight by specifying correct types of stainless. They are strong yet tough, moreover, they resist wear and abrasion.

You get long service life because these alloys are resistant to attack by nearly all oxidizing acid conditions. They assure long, trouble-free performance of equipment, and hygienic cleanliness as well as economy in food, drug, chemical and other process industries.

Leading steel companies produce austenitic chromium-nickel stainless steels in all commercial forms. A list of sources of supply will be furnished on request.

EMBLEM OF INCO SERVICE
TRADE MARK

Over the years, International Nickel has accumulated a fund of useful information on the properties, treatment, fabrication and performance of engineering alloy steels, stainless steels, cast irons, brasses, bronzes, nickel silver, cupro-nickel and other alloys containing nickel. Write for "List A" of available publications.

THE INTERNATIONAL NICKEL COMPANY, INC.
67 WALL STREET
NEW YORK 5, N. Y.

JOURNAL OF Metals

TECHNOLOGY·PRACTICE

EDITORIAL AND BUSINESS STAFF

EDWARD H. ROBIE
Editorial Director

SECTION ONE

ERNEST KIRKENDALL
Editorial and Business Manager
HAROLD A. KNIGHT
Assistant Editor
FRANK T. SISCO
Contributing Editor
GRACE PUGSLEY
Production Manager

SECTION TWO

JOHN V. BEALL
Managing Editor
RUTH L. GARRETT
Assistant Editor
HAROLD N. UNDERHILL
Editorial Assistant

SECTION THREE

E. J. KENNEDY, JR.
Managing Editor
WINIFRED D. GIFFORD
Editor
ERNEST KIRKENDALL
Editor

AIME STAFF

Secretary EDWARD H. ROBIE
Assistant Treasurer H. A. MALONEY
Assistant Secretaries
E. J. KENNEDY, JR.,
ERNEST KIRKENDALL,
WILLIAM H. STRANG (Dallas, Texas)
Assistant to the Secretary
H. NEWELL APPLETON

AIME OFFICERS

President L. E. YOUNG
Vice-Presidents C. H. BENEDICT, A. B. KINZEL, PHILIP KRAFT. D. H. McLAUGHLIN, R. W. THOMAS
V. P. and Treasurer
ANDREW FLETCHER

TECHNICAL PUBLICATIONS COMMITTEE OFFICERS

Chairman E. M. WISE
Vice-Chairmen GEORGE C. HEIKES
GAIL F. MOULTON
Secretary E. J. KENNEDY, JR.
Auxiliary Publications Committee Chairmen (Metals Branch):
Institute of Metals Div.
O. B. J. FRASER
Iron and Steel Div.
MICHAEL TENENBAUM
Extractive Metallurgy Div.
CARLETON C. LONG

●

JOURNAL OF METALS Technology • Practice is on file in many public libraries and is indexed in Engineering Index, Industrial Arts Index, ASM Review and Chemical Abstracts.

Published monthly by the American Institute of Mining and Metallurgical Engineers, Inc., 29 West 39th St., New York 18, N. Y. Domestic subscription, North, South, and Central America, $8; foreign, $9. To AIME members, $6, or $4 in combination with a subscription to *Mining Engineering* or the *Journal of Petroleum Technology*. Single copies, 75 cents. Special issues, $1.50.

Entered as second class matter January 12, 1949, at the post office at New York, N. Y., under the Act of March 3, 1879. Registered cable address, AIME. **Copyright 1949 by the American Institute of Mining and Metallurgical Engineers, Inc.**

Printed in U. S. A.

MAY 1949 Volume 1, No. 5

COVER: Tough pitch copper, partially coarsened by rolling 90 per cent and annealling three minutes at 1000 C. Magnification 5 X. (See Page 312, Transactions section).

section one

EDITORIAL: Buyers' Market Norman Hickman 7
METAL TRACE ELEMENTS IN AGRICULTURE Frank A. Gilbert 8
BLAST FURNACE CONFERENCE REPORT Kurt Neustaetter 12
PROPOSAL TO INCREASE DUES 16
IMPORTANT ACTIONS OF EXECUTIVE COMMITTEES 17
IRON AND STEEL DIVISION DINNER 17
REPORTS OF AIME SECTIONAL MEETINGS 18
ENGINEERING SOCIETIES PERSONNEL SERVICE, INC. 20
PROFESSIONAL SERVICES 20
ADVERTISERS' INDEX 20

section two . . . AIME

To AIME Members 181
Board of Directors' Meeting 182
D. H. McLaughlin for President 184
Where to Split Metallurgists 184
Student Society Reports 185
Local Section Notes 186
Current Institute Papers 187
Calendar of Coming Meetings 187
Drift of Things 188
News of AIME Members 190
Obituaries 197
Membership 198

section three . . . metals transactions

DISCUSSION, INSTITUTE OF METALS DIVISION, 1948 FALL MEETING 297
THE EFFECT OF HIGH COPPER CONTENT ON THE OPERATION OF A LEAD BLAST FURNACE, AND TREATMENT OF THE COPPER AND LEAD PRODUCED A. A. Collins 347

AMERICAN INSTITUTE OF MINING AND METALLURGICAL ENGINEERS, INC.
Metals Branch
AIME also publishes *Mining Engineering* and *Journal of Petroleum Technology*.
Editorial, Advertising and Business Office, 29 West 39th St., New York 18, N. Y.

Convenience and speed of specimen preparation in the METALLURGICAL LABORATORY

First stage of finishing specimens is fine grinding with 600 mesh alundum polishing compound on AB Billiard Cloth. No. 1500-F polishing unit.

Fountains for washing samples during and between polishing.

Second stage of finishing and in many cases the final stage, polishing with AB Metpolish No. 1 or Alumina No. 1 on AB Selvyt Cloth. No. 1500-F polishing unit.

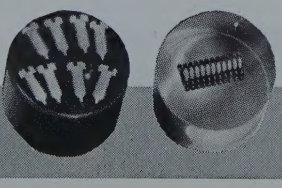
For extra fine polishing, AB Metpolish No. 3 or Alumina No. 3 on AB Miracloth. No. 1505-2F low speed polishing unit.

Buehler

MULTIPLE UNIT POLISHING TABLE

The No. 1540 AB Metallographic Polishing Table with units for three stage polishing is a most convenient piece of laboratory equipment for speedily handling the polishing of specimens. One operator can finish many more specimens in a working day than with a single polishing unit.

The BUEHLER line of specimen preparation equipment includes—
Cut-off Machines • Specimen Mount Presses • Power Grinders
Emery Paper Grinders • Hand Grinders • Belt Surfacers
Polishers • Polishing Cloths • Polishing Abrasives.



Buehler Ltd.
A PARTNERSHIP
METALLURGICAL APPARATUS
165 WEST WACKER DRIVE, CHICAGO 1, ILLINOIS

HEAR YE! HEAR YE!

"One picture is worth a thousand words," said Confucius, or a reasonable facsimile of that gentleman. However, we won't need a thousand words to tell you members about a new cover photo contest—here and now being announced by the JOURNAL OF METALS, your publication.

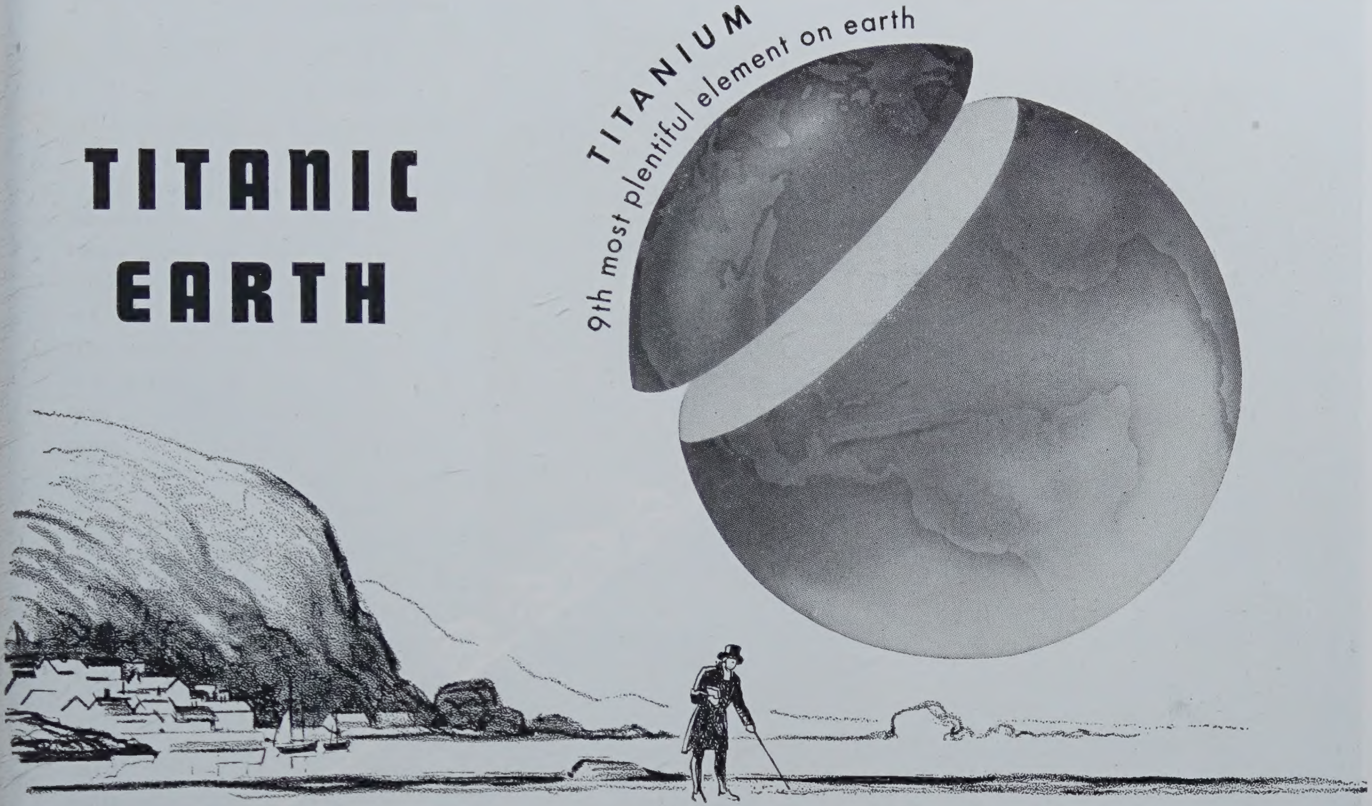
You who are amateur or professional admirers of a good photo—its lights and shadows, its focal points, central figure, blending backgrounds, clear and sharp focus—keep on the lookout for a suitable photo for the front cover of J of M. There are prizes in the offing. It should be related to metals, of course, directly or even remotely, if good enough. The photos should be artistic rather than detailed.

Perhaps you have a photo somewhat bizarre in effect—such as icicles hanging down from a blown-out blast furnace; harvesting prunes with a steel plant in the background (such as might occur in sunny California); perhaps a Republican elephant appearing in a photomicrograph during an election year. Of course, the photo doesn't have to be eerie—just a straight photo in our field, with good artistic points.

The winner will receive a year's dues to AIME for free. Any picture used, if not the first prize winner, will entitle the submitter to any copy of the IMD's symposium volumes. The contest will close Sept. 1, 1949. All entries will be acknowledged by postcard. Glossy prints are essential. All entries are the property of JOURNAL OF METALS. None are returnable.

But come, come, let's have some fun. Let's make this contest a big happy family affair, with good-natured rivalry. Incidentally, it will improve your journal!

TITANIC EARTH


TITANIUM
9th most plentiful element on earth

TITANIUM DISCOVERED

Back in 1791 an English clergyman, William Gregor, who liked to stroll and think on the beaches of Cornwall, became curious about the black sand he saw there. This gentleman of the cloth was also an amateur chemist and in this sand he discovered a new element. Almost coincidentally an Austrian named Heinrich Klaproth (also discoverer of uranium and zirconium) extracted the same thing from rutile and named it "Titanic Earth" for the mythical Titans. Hence our name Titanium.

Thereafter titanium was found in various places including the Ilmen Mountains of Russia (ilmenite) but although it is the ninth element in order of earthly abundance, it remained a mere laboratory curiosity until 1908.

TITANIUM OXIDE

At that time Dr. A. J. Rossi, expert in the reduction of metals, mixed titanium oxide with salad oil to make a white paint. In another 10 years a pure oxide was being produced which quickly won success as a pigment. Paint, false teeth, face powder, tires, shoes, glassware, textiles, inks, plastics, paper consumed an increasing tonnage of titanium oxide but still the pure metal was beyond industry's reach.

TITANIUM METAL & NATIONAL RESEARCH

Titanium is an affectionate metal, over fond of oxygen and nitrogen when at high temperatures. Even a fraction of a per cent of either makes titanium of little value as a structural material. Until recently there was no means of preparing titanium metal in a form sufficiently free of these elements to indicate any potential commercial value. Dr. W. J. Kroll of the Bureau of Mines has initiated many of the recent developments in titanium metallurgy by finding a means of preparing powdered titanium metal.

Only by exclusion of these gases can it be kept from embrittling combinations and when Remington Arms Company, a Du Pont subsidiary, laid its plans to produce metallic titanium in cast and rolled shapes, they knew that at National Research Corporation they could find the knowledge of vacuum technique that they needed.

The melting and casting of titanium was a natural for National Research. We planned the process, designed the equipment and installed it. Today this National Research Corporation pilot equipment is handling the highest quality of commercial metal — not much compared with aluminum — nothing at all compared with steel — but so promising that millions will be spent by the industry within a few years to increase the quantity and lower the price.

USES OF TITANIUM METAL

Titanium stands fourth in abundance among the structural metals and there is plenty in the U. S. A. Tremendous strength, light weight, and remarkable corrosion resistance (comparable only to that of the noble metals) is a unique combination. Coming at a time when long-sighted people are viewing our metallic resources with alarm, it has an assured future. With the price pulled down to a few dollars a pound or less, titanium will be of primary importance to manufacturers of aircraft, automobiles, electric devices, gas turbines, superchargers, marine hardware, rockets, optics, jewelry.

WHAT NEXT?

So, with the help of National Research's high vacuum know-how, another material has been taken from the test tube to the factory. Where else can good men and ideas help — where can they help you? At National Research the best in brains, organization, equipment, and an unequalled accumulation of unique experience are available.

INDUSTRIAL RESEARCH PROCESS DEVELOPMENT
HIGH VACUUM ENGINEERING & EQUIPMENT
Metallurgy — Dehydration — Distillation — Coating — Applied Physics

NATIONAL RESEARCH CORPORATION
SEVENTY MEMORIAL DRIVE CAMBRIDGE, MASSACHUSETTS

In the United Kingdom, BRITISH-AMERICAN RESEARCH, LTD., London S.W 7, England—Glasgow S.W 2, Scotland

The rapid advance of modern iron and steel-making has been due not only to improved use of standard materials but also to the introduction of new ones.

Boron has established itself definitely among the resources of iron and steel manufacture. In the economical and easily soluble form of Ferro-Boron it is approved and used by leading producers to heighten the qualities of cast iron, malleable iron, cast steel, rolled steel, tool steel and other ferrous products.

Ferro-Boron is available for all demands. In the United States, it has the advantage of abundant raw materials found within our own borders. The Molybdenum Corporation, a large producer of Boron alloys and a pioneer in developing their uses, will welcome your inquiries on any application of Molybdenum, Tungsten, or Boron.

MCA

AMERICAN Production, American Distribution, American Control, Completely Integrated.

Offices: Pittsburgh, New York, Chicago, Cleveland, Detroit, Los Angeles, San Francisco, Seattle.

Sales Representatives: Edgar L. Fink, Detroit; Brumley-Donaldson Co., Los Angeles, San Francisco, Seattle.

Subsidiaries: Cleveland-Tungsten, Inc., Cleveland, O.; General Tungsten Manufacturing Co., Inc., Union City, N. J.

Works: Washington, Pa.; York, Pa.

Mines: Questa, New Mexico; Urad, Colorado.

MOLYBDENUM CORPORATION OF AMERICA
GRANT BUILDING PITTSBURGH, PA.

MCA

JOURNAL OF
Metals
TECHNOLOGY·PRACTICE

GUEST EDITORIAL

NORMAN HICKMAN • VICE PRESIDENT, THE AMERICAN METAL COMPANY, LTD.

BUYERS' MARKET

The era of shortages, of controlled and, then, rising prices seems to have ended. Many of us, concerned with the distribution and sale of non-ferrous metals, find ourselves, as it were, in a strange country with no familiar landmarks to guide us. Many who have matured during the past decade in this important branch of the metal industry have become adept in varying degree in expressing regretful inability to meet exigent demand. A few have even learned to say "so sorry" with compelling sincerity. It is a bit of a shock to have these words adopted so quickly and so universally by our consuming friends. To a few of us, those who remember the early twenties and the early thirties, it is another variant on the old theme: "This is where we came in."

However, there is a distinct difference between the current buyers' market and those of the past—not a complete difference, for improved practices grow gradually. But there is today, in the writer's judgment, a sounder and franker relationship between buyer and seller than existed a generation ago. The complementary nature of the one to the other is more generally recognized. Fortifying this trend is the more complete and up-to-date statistical data which is increasingly being made available both here and abroad on an industry wide basis, i.e., from the initial stages of raw material production to the outgo of primary fabricated products. The latest and most complete flow sheet of this kind is that published by the British Bureau of Non-Ferrous Metals, which is distinguished by the immediacy of its figures. It might be injected that this is work of a private, not of a government, bureau.

Phrases such as "service" and "sales engineering," though sadly overworked and often misused, indicate, however, another change that has developed between the producers and consumers of non-ferrous metals. The outstanding achievement of the International Nickel Company in broadening the uses of its product is a pointed example of the trend toward precise information redounding to the mutual benefit of buyer and seller. In varying degree the trend is universal in the metal field. Raw material producers and consumers find an increasing challenge to their problem of producing and specifying materials able to withstand more exacting initial processing and ultimate use.

Today's buyers' market throws important responsibility on consumers and emphasizes benefits which some of us believe can be derived from a more complete confidence between buyer and seller. For example, inventories are being drastically reduced. Ultimate consumptive use, however, has in the case of many non-ferrous metals, shown no such comparative drop. In interpreting this situation proper cooperation between buying and selling organizations can inspire policies based on facts, not on fears. Once the latter dominates, industrialists, whether immediately concerned with buying or selling, jointly justify the taunt of the enemies of our system that ours is a "boom or bust" economy. Such an economy can cumulatively stem from deliberate concealment of the truth by one branch of business from another. Buyers and sellers alike should pool information so that both imaginary shortages and imaginary surpluses are minimized. There are, unfortunately, still with us those few who aim to gain a short-range and illusionary profit by failure to realize that sound buying and selling today should adopt a policy in which is merged two slogans, one quite old and one relatively new: "Truth will out" and "Eventually, why not now?"

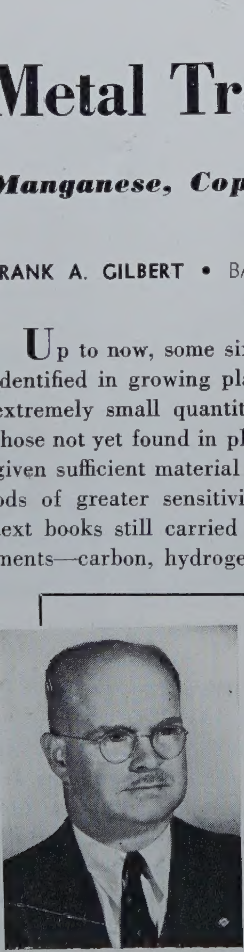
Tobacco that has received fertilizer only. It is inferior only by comparison with field on opposite page.

Metal Trace Elements in Agriculture

Manganese, Copper, Zinc, Boron, "Moly", Cobalt, Etc.

FRANK A. GILBERT • BATTELLE MEMORIAL INSTITUTE

Up to now, some sixty elements have been positively identified in growing plants, most of these, of course, in extremely small quantities. It is possible that many of those not yet found in plants may eventually be identified, given sufficient material for analysis and analytical methods of greater sensitivity. Twenty-five years ago some text books still carried the statement that only ten elements—carbon, hydrogen, oxygen, nitrogen, phosphorus, potassium, sulphur, calcium, magnesium, and iron—were essential to plants; and that these with the addition of sodium, chlorine, and iodine were all that were necessary for animal existence.



Frank A. Gilbert is a research botanist on the agricultural science staff at Battelle Institute. He has been closely identified with most of the research at Battelle on the effects of metals and chemicals in agriculture and is author of the book, "Mineral Nutrition of Plants and Animals," published in 1948.

A graduate of Massachusetts Agricultural College, Dr. Gilbert received his Master of Arts and Doctor of Philosophy degrees from Harvard University. He was at one time a plant pathologist with the U. S. Department of Agriculture and served as professor of botany at Marshall College, Huntington, W. Va., for fifteen years. He joined the Battelle staff in 1944.

Dr. Gilbert is a member of the Botanical Society of America, the Mycological Society of America, the American Association for the Advancement of Science, and Sigma Xi honor society. He has published over two-score scientific papers in the fields of plant nutrition, mycology, and botany.

Today we know that at least six additional elements are required, either by plants, by animals, or by both: Manganese, copper, zinc, boron, molybdenum, and cobalt —which along with iron and iodine, already mentioned, are called trace elements, magnesium and sulphur frequently being included in this category. The term trace element, or preferably "micronutrient," is applied because of the minute amount found in the plant or animal body and has nothing to do with its value to the organism, which is out of proportion to the relative quantity present.

Early investigators did not discover the true worth of micronutrients although their presence in plant and animal ash was demonstrated along with other elements, at present of no proven value. Refinement of technique had not been developed to the point where nutrient solutions or diets could be purified sufficiently to be free from traces of the element under test.

However, by means of improved laboratory technique, manganese was proved, in 1922, to be essential to plant life, zinc and boron in 1926, copper in 1931, and molybdenum in 1939. For animals, copper was discovered to be necessary in conjunction with iron for hemoglobin formation, as a result of a series of studies beginning in 1925. Manganese was not considered as essential until about 1932, while the necessity for zinc in the diet was proved in 1935. The essentiality of cobalt was also demonstrated that year although its need had been suspected previously. Magnesium had been known to be a plant nutrient since the last century, but our knowledge of its necessity to animals dates only from 1932. It may be

Tobacco treated with copper sulphate in addition to fertilizer. Note that the leaves are larger and healthier.

seen, therefore, that with the exception of iodine and sulphur, our understanding of the value of trace elements is quite recent and the volume of literature on the subject has accumulated largely during the past ten years.

The soils of this country were not equally productive even before the settlement of the pioneers who wrested the land from the aborigines. When a homestead was to be carved out of the forest and a site was to be had for the choosing, the prospective farmer preferred land on which walnut and hickory grew to that on which beech was dominant. He had learned that certain trees were characteristic of the deeper, more fertile soils, and located his home accordingly. However, practically all land, even that of limited natural mineral content, would produce fair crops provided that the buffering organic matter and nutrient materials present were returned to the soil where they could be used again in the natural cycle of growth and decay. This cycle had gone on for millions of years, and the occasional corn patch that the aborigines scratched out in some stream bottom, where the virgin forest did not interfere too greatly, had little effect on the balance of nature.

man squandered resources

Civilized man, with his usual ardor for exploiting natural resources, did not behave differently in America. The forests were cut down and the rich deposit beneath was despoiled. The poorer soils were naturally the first to feel the effects of misuse. When the original settler who had "improved" the country by cutting trees, building trails, and making further despoilation easier, moved on to carry out his destructive work farther West, the naked land was left unprotected, to be further injured by erosion and leaching.

The better soils took longer to ruin, but eventually, many of these also succumbed, insofar as their ability to produce a good crop was concerned, and were abandoned. The race to ruin more land as fast as possible, eventually slowed down. The Pacific was reached and, like waves beating against a rock, the backwash of civilization scattered out to settle on locations that were thought worthy of salvage or were overlooked in the first Westward rush.

Nature is a great healer and, in cases where abandoned land was not too badly eroded, the scars were covered by protective green growth which undertook the extremely slow process of again building up a productive soil. The land was injured or ruined for farming, however, for many generations.

The first elements requisite for plant growth, which became depleted in the overworked soil, were those most soluble or needed in greatest abundance. The supply of trace elements, required only in small amounts, but without which plants could not grow satisfactorily, took much longer to deplete. Formerly, the need for them was lost sight of in view of the requirement for greater amounts of nitrogen, phosphorus, potash, and calcium.

As soils became depleted, the application of animal manures or organic fertilizers, such as fish meal, became more or less general. Organic materials of this sort are rich in the micronutrients, including copper, zinc, and manganese, necessary for the growth of a normal crop. So long as organic fertilizers were used there was no micronutrient problem.

The low-analysis fertilizers, usually containing at least nitrogen in organic form, also supplied enough micronutrients to satisfy plant needs on most soils. Today, however, animal manures are scarce. The high-analysis, purified, and frequently synthetic fertilizers contain micronutrients as traces only, usually not nearly enough to satisfy plant needs. As a result, lack of such necessities for balanced plant and animal nutrition is becoming more and more evident, especially on soils not naturally rich in these elements. Soils subject to severe leaching are also prone to be deficient in trace elements. Thus the Atlantic coastal plain, which has been farmed for several hundred years, and which is sandy and easily leached, seems to show great need for several of the micronutrients.

specific deficiencies

Cobalt deficiencies have been reported from some of the coastal plain states as well as from Wisconsin and Michigan. In some cases in these areas, it has been difficult to raise cattle and sheep because of this deficiency in the forage grown on these soils. Copper and magnesium deficiencies are more or less general along the coast, and even soil areas farther inland frequently respond to copper treatments. Soils with a high organic content almost always are improved by the addition of copper, the classic example being the soil of the saw-grass region of Florida where it was found impossible to raise vegetables unless the land had previously received a copper amendment.

Zinc deficiencies of citrus and other crops occur in Florida and adjacent states and also in California where good growth of citrus in some locations has become nearly impossible, unless the trees receive zinc sprays, or the element in some other form. In the Pacific Northwest, certain areas are low in sulphur.

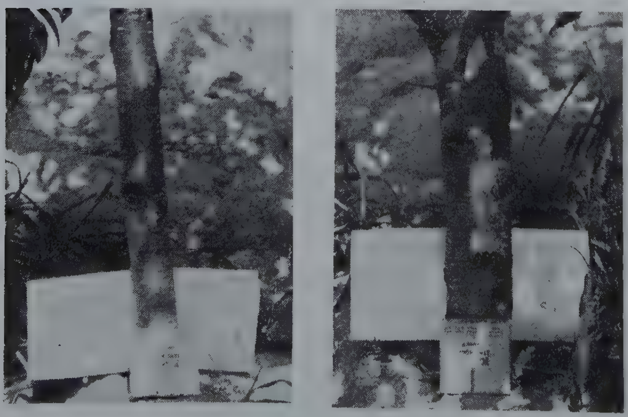
Although most soils contain large amounts of iron, even deficiencies of this element have been recorded from Florida and Massachusetts. Manganese and boron deficiencies are widely scattered, but are even more common than those previously mentioned. In fact, boron deficiency has been found to be so general that the treatment of alfalfa fields with this element has become a common practice. Even in the case of molybdenum, which is almost universally present in soils, it has been found advantageous to include a small amount in the fertilizer to be used in some areas.

The need of iodized salt for human and animal consumption in the so-called goiter belt is well known. The soil of this belt, which extends from the Pacific Northwest across the United States through Colorado, the Dakotas, Iowa, and the Great Lakes states, is naturally low in iodine, so the incidence of goiter in the area has been much higher than in the states to the South and West, where the land has been more recently submerged beneath the sea.

plants and animals affected

The more severe trace-element deficiencies make their appearance as physiological diseases of plants or of animals. In the case of magnesium deficiency of tobacco, a disease known as "Sand Drown" occurs in which the plant assumes a characteristic chlorotic condition; and in severe cases, all growth stops. A lack of sulphur or of manganese also shows up in plants as a chlorosis, characteristic in each case. The "Gray Speck" disease of oats and the "Marsh Spot" of peas have been definitely proved to be due to a deficiency of manganese. When this element is low in the animal diet, sterility tends to occur, and a disease of chickens known as "Slip Tendon" has been found to be associated with too little manganese, along with insufficient choline, a vitamin of the B complex.

A lack of copper in citrus soils results in "Dieback," while the term "Reclamation" disease is commonly used in Europe for lack of growth of any crop on newly cultivated lands deficient in this element. There have been several deficiency diseases of livestock associated with low-copper soils and forage. More than a decade before the role of copper in hemoglobin formation was discovered, investigators in Northern Europe were studying possible mineral deficiencies in the forage in areas where a wasting disease of cattle called "Lechsucht" was common. In 1933, its cause was established as a copper deficiency by the finding of marked differences in the copper content of the forage in healthy and sick areas and by the curing of the trouble with copper therapy.



Trunk of peach tree that received no trace elements (at left); trunk of tree that received trace elements (at right).

It has also been established that a disease of lambs, called "enzootic ataxia," in which nervous symptoms occur, is caused by a deficiency of copper in the ration of pregnant ewes. In Australia, severe cases of copper deficiency cause the "falling disease" characterized by staggering, falling, and eventual death. A lesser deficiency results in the development of "steely" or "stringy" wool where a progressive change takes place from a normal to a straight, glistening type of fiber. Attempts to breed sheep with this kind of wool result in the "falling disease" or ataxia in the offspring.

One of the most interesting chapters in the history of our knowledge of the use of trace elements has been in the discovery of the value of cobalt. This element is not essential to plants, but is usually taken up in small amounts. Animals, under natural conditions, are compelled to obtain it from the forage on which they feed. The occurrence of deficiency diseases of sheep and cattle characterized by anemia, and quite different from the ataxia caused by the lack of copper, is widespread throughout the world. An early and still useful treatment was to shift the animals to pastures known to be healthy or to add crude iron salts (limonite) to the feed. As this resulted in a cure, it was formerly believed that the anemic conditions were due entirely to a lack of iron. When pure iron preparations were used, however, the animals did not always respond. Research has shown that, in such cases, cobalt is usually the missing element.

In the United States, cobalt deficiency in livestock has been reported from New Hampshire, New York, Massachusetts, North Carolina, and Florida, in addition to the lake areas of Michigan and Wisconsin, where the trouble is called "Grand Traverse" or "Lakeshore" disease. The requirements per unit of body weight are apparently higher for ruminants than for other herbivores since horses can thrive on pastures where cattle and sheep develop serious trouble.

The amount of cobalt necessary for recovery is remarkably low, and sheep will return to normal with daily intakes as low as 0.1 mg. It has also been found that two pounds of cobaltous sulphate per acre are sufficient to restore the pasturage on deficient soils to a healthy level.

most research with boron

More research has been done on boron in the past few years than on any other trace element. The amount necessary for the optimum functioning of plants is relatively low, and the range between deficiency and toxicity is narrow. Soils, which under ordinary conditions will furnish enough boron for most plants, may become deficient when the pH is raised to an unusually high level, as sometimes occurs through overliming. At present, the fertilizer value of boron is fully appreciated, and a deficiency of it is recognized as the causative agent for "heart rot" of beets, "brown heart" of turnips, "cracked stem" of celery, "top sickness" of tobacco, "internal cork" of apples, "browning" of cauliflower, and "yellows" of alfalfa.

With an increasing awareness of the value of trace elements in plant and animal nutrition, the question arises of what to do about it. Extreme deficiencies, when the cause is known, are soon corrected. Borax is added to the alfalfa field; sick citrus trees are sprayed with a zinc salt; copper sulphate is applied to peat soils on which vegetables are to be grown.

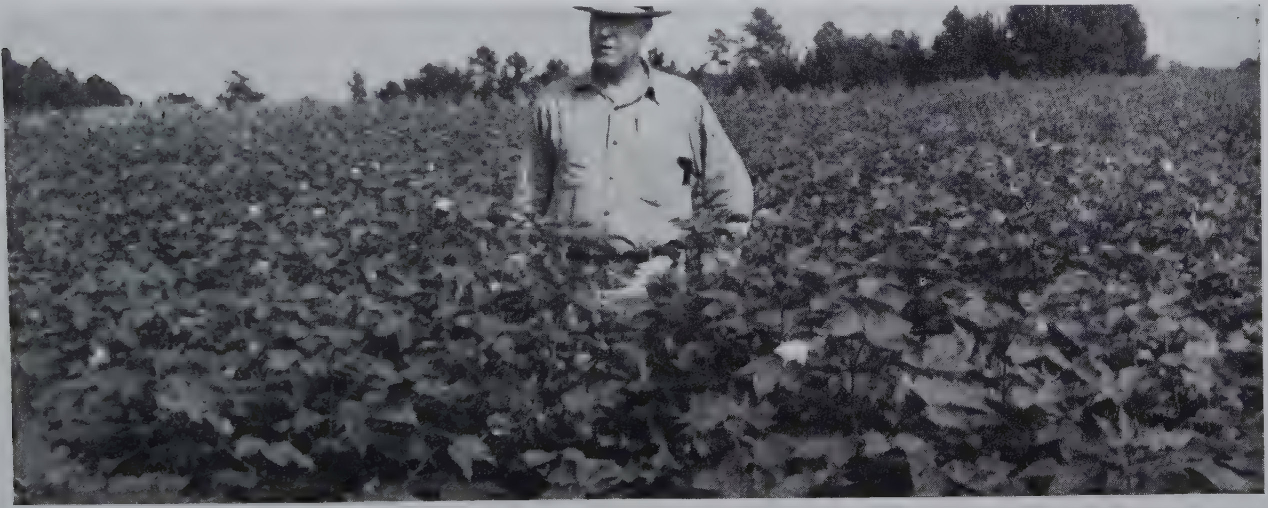
case history: peach trees

However, hundreds of borderline cases occur in which a retarded growth is the only symptom, and the crops and animals do poorly but pass as normal. Dozens of cases may be cited, but examples will be given of only two. The sand-hills region of the Carolinas is one of the peach growing areas of the United States. The trees are heavily fertilized, but in general little thought has been given to applications of trace elements. Where applications have been made, however, the results have been remarkable. The trees are larger, greener, and more vigorous, and greatly outyield both the control trees on the test farms and the trees on nearby farms which pass as normal.

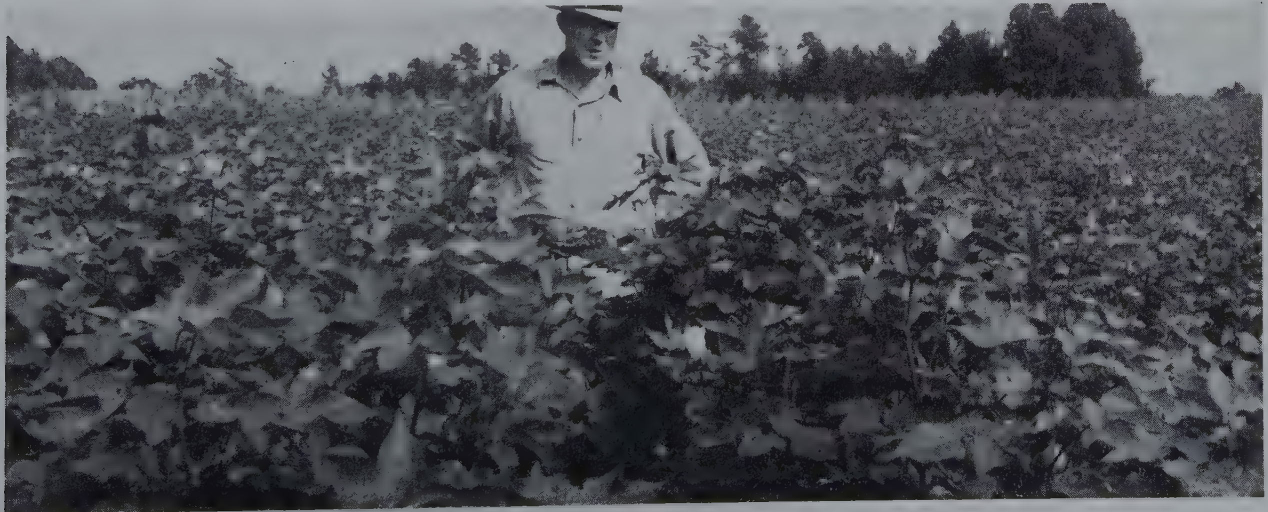
In several of the tobacco growing regions, extensive tests have been carried out in the past few years on the value of copper as a soil amendment. Although none of the plants in the untreated fields showed any sign of copper deficiency, the growers in 80 per cent of the tests obtained increases in weight, quality, or both for their treated tobacco. Other crops have shown a similar response.

Fertilizer, fortified with the trace elements, is being sold more and more frequently, and the demand is growing. At present, the amounts of copper, zinc, manganese, and boron going into fertilizer and feed amendments is very small as compared to the amounts devoted to other uses.

However, taking into consideration that the total amount of fertilizer sold during 1947 was over fifteen million tons and that increasing percentages in the coming years will be fortified with trace elements, the future use for this purpose looks very promising.



Cotton here has received fertilizer treatment only. It is satisfactory except when compared with below.



This cotton was treated with copper sulphate amendment, besides fertilizer. Leaves and stems are larger.

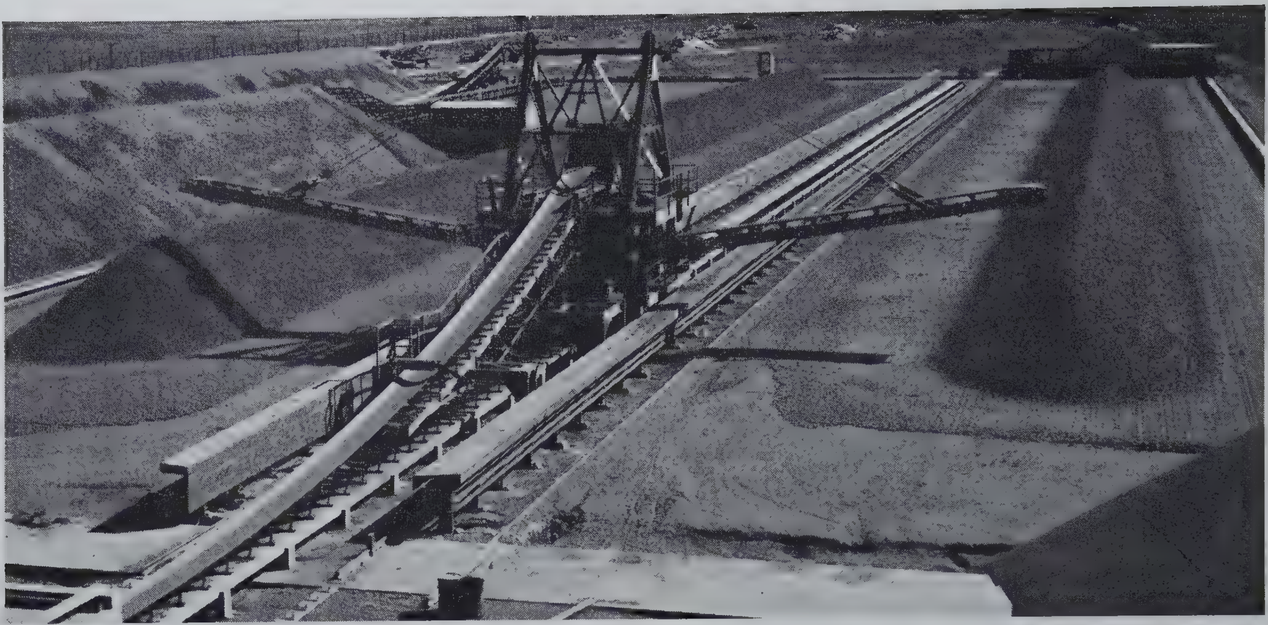


Fig. 2—A double wing stacker and composite unit deposit many thin layers of ore over one another. Each bed holds 17,000 net tons of ore when filled.

Blast Furnace, Coke Oven and Raw Materials Conference

Great Progress Made at Far Western Blast Furnaces

KURT NEUSTAETTER • BLAST FURNACE ENGINEER, INLAND STEEL CO.

In a recent issue of one of this country's leading magazines, blast furnace men were accused of having made little progress during the last 75 years. It is unlikely that anyone out of the 330 operators, maintenance and construction men who attended the Blast Furnace, Coke Oven and Raw Materials Committee meeting of the A.I.M.E. at Chicago from April 18-20 will share this opinion.

For many years the growth of the steel industry west of the Mississippi had been slow. As Edward J. Duffy of the Kaiser & Frazer Parts Corp., Ironton, Utah, pointed out, the greatest obstacle to the early erection of a fully integrated steel plant in the Far West was the false rumor that there were no iron bodies large enough or of proper chemical analysis to support blast furnaces.

With the rapid development of industry in the West the absence of a strong basic steel industry became more and more of a challenge which had to be met. As Toynbee points out, growth of civilization is characterized by successful response to a challenge brought on by changed conditions. The West has responded splendidly.

blending ore

Drillings of the Bureau of Mines have proven that large ore bodies are available and the blending processes developed by Columbia Steel Corp., Kaiser Co., Inc., Geneva Steel Co. and the Colorado Fuel & Iron Corp. have proven that the ore can be used successfully.

The blending operations practised by above companies were described by Edward J. Duffy, Robert R. Williams, Jr., of Colorado Fuel & Iron Corp. and C. L. Waggoner, Geneva Steel Co. These people have not just solved their local problems, they have gone new ways and they have actually succeeded in filling the old established Easterners and Midwesterners with envy.

Kaiser operates three ore properties at Iron Springs, Utah, the Armstrong, Smith and Lindsay claims. All three deposits vary greatly in their analyses. The Armstrong and Smith properties consist up to 80 per cent magnetite, associated with some hematite and very little limonite. The Lindsay ore body consists up to 40 per cent magnetite, associated with greater amounts of hematite and limonite. On all three properties iron runs from 50 to 60 per cent, phosphorous from 0.106 to 0.618, sulphur and manganese are low, silica runs from 0.3 to 18.00.

These deposits are mined by the open pit method. The ore bodies dip generally in excess of 15 deg, requiring removal of some overburden.

In operating the mines, core drillings are taken at intervals of 100 ft and samples are also taken from the wagon drill tailings at intervals of 25 ft. From the assays of samples thus obtained a map is maintained on the working face of the mine, showing daily analyses of the phosphorous and silica content of the ore. Ore in excess of 0.50 phosphorous is set aside, ore below 0.50 phosphorous is crushed and reduced to minus 10 in. The crushed ore is conveyed to two stock piles containing 0.36 to 0.50 and 0.35 and under phosphorous ore. Low phosphorous ore is reclaimed by a reclaiming conveyor belt and carried to a double deck vibrating screen. One deck is 5 in. and the second is 2½ in. The plus 5 in. and plus 2½ in. material passes through a secondary crusher set at 2½ in. All minus 2½ in. ore is then carried on a conveyor belt to a bedding stock pile. The bedding system consists of:

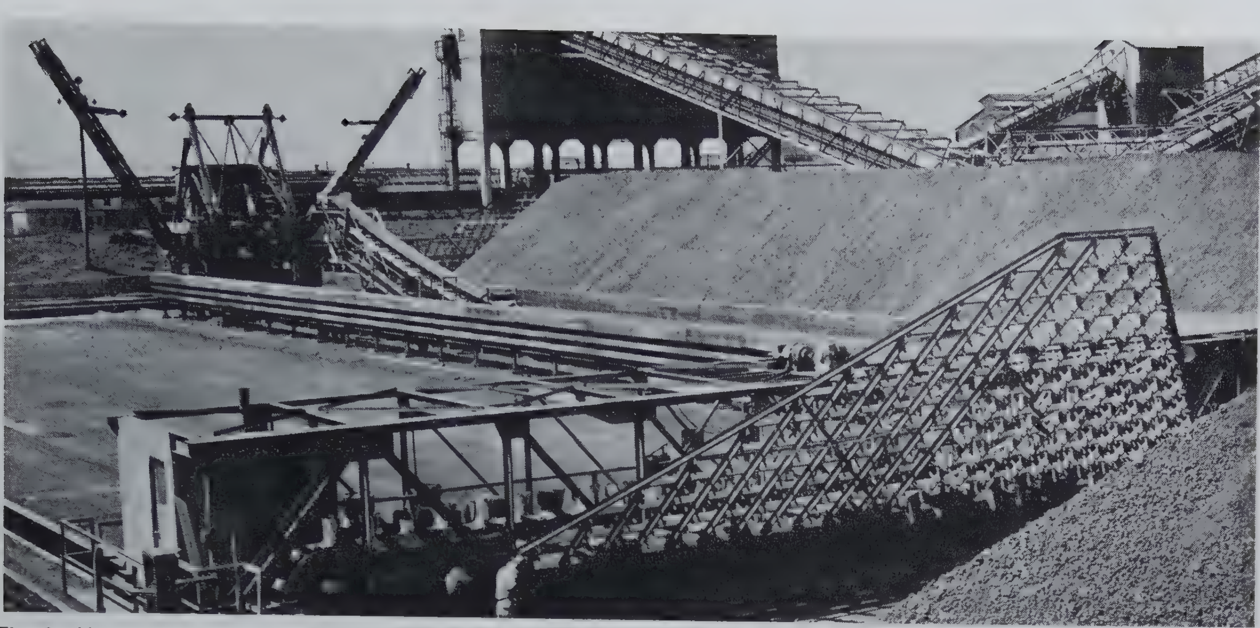


Fig. 3—Here is a Robins-Messiter reclaiming machine, used at the Geneva Steel Co. It is essentially a traveling bridge, spanning the width of the ore bed and a trench conveyer.

368 ft. highline, 75 ft. high.

1 Robbins traveling tripper, speed 20 ft. per min.

368 ft. reclaiming tunnel and conveyor belt under the pile.

2 Smith-Adams variable position feeders.

18 Chutes, spaced at 20 ft. centers, openings 25 x 25 in.

18 Air operated ore gates.

Size and capacity of the bedding system:

Length of bed, 368 ft.

Height of bed, 75 ft.

Incline angle of bed, 40 deg.

Capacity, 214,000 net tons.

From the reclaiming conveyor the material can be separated by another double deck vibrating screen equipped with 2 and ⅝ in. screens. The ore is loaded into railroad cars and shipped to either Fontana or Ironton. A similar operation is carried out by Kaiser at the Eagle Mountain Mine in California.

Mr. Duffy then showed a most interesting slide, see Fig. 1. The slide illustrated the beneficiation of bedded ore compared with run-of-mine ore. It should be noted from the graph that silica variation over the period in question was only 1 per cent. For the same period run-of-mine ore had a silica variation of 17.60 per cent. The spectacular improvement of furnace operation due to blending and sintering at Fontana was reported at last year's meeting by Mr. Saussaman.

Mr. Williams described the ore preparation plant in operation at Pueblo since 1943. The Pueblo furnaces receive ore mainly from the Sunrise Mine in Wyoming and the Iron Mountain Region in Utah. The ores show wide variations in chemical content, accompanied by irregularities in sizing. These variations were thrown directly on the furnaces, and the effects were that many burden changes were required and the pig iron produced reflected variations that could not be corrected by operative changes. Design work was begun in 1941 and construction work was completed in September, 1943, on an ore preparation plant consisting of crushing and screening units, sintering facilities and blending beds. Due to space requirements the site of the plant is about 1¼ miles from the blast furnaces.

The crushing and screening plant produces minus ¼ in. ore for sintering and minus 2 in. plus ¼ in. material for the bedding system. Approximately 700 tons of ore per hour can be treated.

There are four beds on which ore may be blended. The normal practice is to use two beds for stacking while reclaiming from the two remaining beds. A double wing stacker (see Fig. 2) is coupled to a trailer tripper, and the composite unit is traversed at speeds of 60 to 110 ft. per min. This traverse is back and forth along the beds, with automatic reversal of direction. Thus, many thin layers of ore, one over the other, are placed on the beds to form a pile triangular in cross section, about 50 ft. at the base, 18 ft. in height and 560 ft. in length. Each bed will hold approximately 17,000 net tons of ore when filled.

The ore is reclaimed by a Robins-Messiter reclaiming machine (Fig. 3 shows a similar reclaimer used at Geneva). The machine is essentially a traveling bridge, 75 ft. center to center of rails, spanning the width of the ore bed and a trench conveyor. A harrow, adjustable for slope and triangular in shape to cover the cross sec-

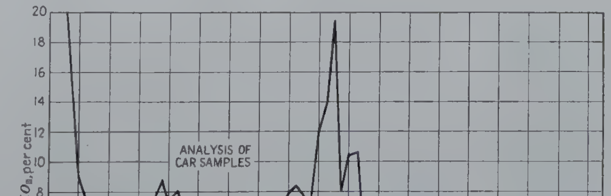


Fig. 1—Bedded ore has silica variation of only 1 per cent vs. 17.60 per cent for run-of-mine ore.

tion of the bed, is mounted on the front end of the bridge structure; it is a structural steel grid frame with vertical and horizontal members spaced about 2 ft. apart. The vertical members are provided with harrow teeth closely spaced. A reciprocating motion is imparted to the harrow by means of a connecting rod and crank. A reversible plow conveyor is suspended under the forward side of the main bridge structure, consisting of a single strand of steel bar chain to which are attached the plows. The lower strand of the chain operates in a steel trough to carry the dislodged ore to one or to the other end of the machine for discharge to a trench conveyor. Subsequently it is carried by two belts to the ore loading station where it is loaded in railroad cars for the furnace bins. The capacity of the system handling ore reclaimed from the beds is 800 net tons per hour.

Extensive weighing and sampling units are available. During a two-week period in November, 1948, a total of 31,641 net tons of ore was stacked in beds Nos. 1 and 2. The iron content of the ore varied from 46.2 to 63.9 per cent, silica from 5.0 to 14.3 per cent, alumina 0.20 to 7.50 per cent, lime from 1.00 to 3.80 per cent, magnesia from 0.30 to 0.70 per cent, phosphorous from 0.067 to 0.130 per cent, sulphur from nil to 0.04 per cent.

The calculated number of round trips the stacker made in stacking the beds was 480. There were, then, 960 layers in each of the beds which were cut through in nearly a vertical plane in reclaiming the ore.

When reclaiming the ore the iron variation had dropped to 51.5 to 55.4 per cent, silica 9.7 to 12.7 per cent, alumina 2.3 to 3.8 per cent, lime 1.7 to 3.1 per cent, magnesia 0.40 to 0.53 per cent, phosphorous 0.088 to 0.110 per cent, sulphur 0.004 to 0.008 per cent.

Statistical analysis of the characteristics of the blended ore showed excellent distribution. Frequency diagrams were not skewed, and the five separate groups of raw material going to the beds had been so blended as to appear as a single group in the frequency diagram.

Furnace operating data presented by Mr. Williams showed that the daily average variation of silicon in the iron dropped from 0.57 to 0.25 per cent, the sulphur variation dropped from 0.019 to 0.013 per cent. The total number of burden changes necessary per month dropped from 97 to 22.

Mr. Waggoner reported about the ore blending at Geneva, pointing out that Lake Superior ore is mixed several times in the loading chutes, in the loading and unloading of railroad cars and ore ships. Many faces are worked at the mines in the Superior region and the ore goes to many consumers. By comparison, there is just one loading of Geneva ore at the mine, one unloading at the plant, with no blending during transportation. Seasonal transportation limitations necessitate large stock piles of Lake Superior ore, with a consequent further blending occurring as the ore is spread and later picked up by the ore bridges. Great Lakes operators receive several grades of ore; only one grade is received at Geneva where an efficient ore blending operation is a necessity.

The need for efficiency in Geneva's bedding operation is apparent when it is realized that, on the present rate of three furnace ore consumption, it is necessary to make complete burden changes on all furnaces every four or five days. Such changes, often radical, are necessary since the area of the bedding piles is limited.

Mining and crushing operations are similar to the ones at Iron Springs. The blending operation is very similar to the one practiced at Pueblo. Robin's stackers and reclaimers are used. There are four pairs of piles averaging 13,000 tons per pile, giving a total bedding capacity of 104,000 tons. The piles average 58 ft. wide, 21½ ft. high and 360 ft. long.

Mr. Waggoner showed a number of charts and tables showing that sister piles can be considered as having much the same analysis and being nearly identical to the mine sample average. Thus, mine sample results can be used for furnace burden calculations with no appreciable errors in furnace operation. No better proof for the efficiency of the blending could be asked for.

The blending session however was only one of the highlights of the meeting.

coke quality

J. F. Peters, Inland Steel Co., who presented a paper on "The Effect of Coke Stability on the Operation of the Blast Furnace," is one blast furnace man who does not blame his troubles on the coke without some well substantiated data. Mr. Peters has been operating two blast furnaces constructed during the war for several years. The plant does not have coal resources of its own and a large variety of coals has to be bought on the open market; he therefore was subjected to ups and downs of coke quality more than most other operators.

Mr. Peters pointed out that size and analysis are only two of the many qualities of coke and that the value of coke for blast furnaces is determined by the sum of all its qualities. He has studied the stability of coke, has correlated it with the amount of braize screened out of the coke and in turn correlated this data with blast furnace operation.

The undersize coke at Plant No. 3 of Inland Steel Co. is separated in three sizes, minus 1½ in., minus 1 in., and minus ⅝ in. The amount of the minus ⅝ in. braize in comparison with the total amount of coke produced proved to be significant for blast furnace operation.

In addition to providing heat and reducing power, coke must be able to stand the crushing effect of the stock column in order to permit a smooth flow of gases. If the coke fails to withstand crushing, the gas flow throughout the stack is restricted and irregular. Mr. Peters defined "stability" of coke as its ability to withstand crushing in the furnace. He showed that amount of braize produced and the ASTM tumbler test are of equal value as a measurement of coke stability.

A large number of examples were shown backing up this contention.

For the month of May, 1945, the following data were given for "B" Furnace:

	% Braize	*% Vol. Matter*	*ASTM Stability*	*Av. Daily Tonnage*
May 1-21	6.90	.652	42.0	1206
May 22-31	8.30	.810	39.5	1150

Period of Sept. 1 to Nov. 10, 1948: The braize was averaging 6.66 per cent from Sept. 1 to 16, "A" and "B" furnace averaged 1155 and 1127 tons per day respectively; the furnaces were operating normally with no evidence of hanging or slipping. Starting Sept. 16, the braize started to increase, averaging 8.12 per cent for the period of Sept. 16 to Oct. 15, an increase of 22 per cent; the tonnage produced dropped to 1058 and 1035 tons respectively, decreases of 97 and 92 tons per day. The wind had to be curtailed during this period to reduce blast pressure and to prevent the furnace from hanging excessively. Irregular gas flow was noted on the inwall temperatures, indicating poor peripheral gas distribution. For the period Oct. 16 to Nov. 10 the braize production decreased to 6.20 per cent, the production increased 72 tons per day on both furnaces.

Mr. Peters also showed that the importance of stability is far greater than the importance of ash content. He compared two months of operation, one with 11.87 per cent ash, 7.7 per cent braize and a tonnage of 1090 tons and one with 10.55 per cent ash, 9.5 per cent braize and a tonnage of 1014 tons.

small vs large furnaces

A few of the most prominent blast furnace men of the country discussed the relative merits of small and large blast furnaces. G. P. Burks, Gary Works, Carnegie-Illinois Corp., showed that the trend toward larger furnaces has been strong during the last 15 years. While in 1935 20.3 per cent of all furnaces in the United States had 15 ft. or less hearth diameter, in 1948 only 9.1 per cent of all furnaces were so small; during the same time the number of furnaces 25 ft. and over increased from 2.3 per cent of the total to 24.1 per cent of the total.

Mr. Burks compared the performance of five 20 ft. 6 in. furnaces with the performance of five furnaces of 25 ft. to 26 ft. 6 in. hearth diameter for the period of 1942 to 1947. He posed three questions.

Question No. 1: Does the quality of the raw materials effect one type of furnace more than it does the other? He found that with decreasing quality of raw materials, the iron production decreased regardless of the size of the furnace, and that with decreasing quality of raw materials the iron production on the large furnaces decreased more, proportionately, than it did on the small furnaces. The wind rate on the large furnaces dropped more rapidly than it did on the small furnaces. Since operating conditions and methods were comparable and since the maximum wind, short of overblowing, was blown in each case, it is logical to assume that the decrease in wind rate was a resultant of deteriorating raw material quality. As the latter became progressively worse, with attendant increasing roughness of operation, it becomes mandatory to reduce wind until furnace control was attained and flue dust production was within reasonable limits.

Question No. 2: Which type of furnace has the lower coke rate? The curves obtained by Mr. Burks showed variations but the trend lines indicated that the coke rates were practically parallel with the small furnaces showing an average of 44 pounds of coke less per ton of iron. However, Mr. Burks claimed that this difference was due to different design of the furnaces and he comes to the conclusion that there is no appreciable difference in coke consumption between large and small furnaces.

Question No. 3: Is there a difference in the uniformity of the product in a comparison of large and small furnaces? A study of standard deviations of silicon content showed that the small furnaces deviated more in analysis than the large ones. This might be expected when it is reasoned that the large furnaces hold a greater amount of iron in the hearth and therefore there is more mixing from cast to cast.

J. Knox Dye of the Ashland Division, Armco Steel Corp., selected four items of comparison for two small furnaces of 13 ft. 9 in. and 14 ft. 6 in. hearth diameters and one large furnace of 25 ft. 9 in. hearth diameter. Investigation of over 4600 casts on each of the three furnaces showed that the chemical quality of production requires more operating control as the size of the furnaces increases. The standard deviations of silicon increased with increasing furnace size.

Mr. Dye found a decided difference between large and small furnaces when taking off the wind. Extreme care must be taken when coming down with the wind on the large furnace. This is true even after cast when taking off the wind to change a tuyere or monkey; even then there is the danger of filling a blow pipe. Mr. Dye blames the difference on weak coke. The coke breaks badly in the large furnace and the hearth is usually filled with small coke which prevents the slag and iron from draining freely from the tuyere area.

Mr. Dye also pointed out that higher grade materials are necessary for dams, linings of runners and tap hole clay. During a four year period the small furnaces, 8 tuyeres each, lost 201 tuyeres. while the large furnace, 16 tuyeres, lost 292 tuyeres; the small furnaces lost 25 and 37 monkeys respectively, while the large furnace lost 47; there were 8 coolers against 26, but 32 bosh plates against 2. These figures indicate that the loss of tuyeres, monkeys and coolers varies directly with the furnace size, while the loss of bosh plates varies inversely with the furnace size.

C. I. Hoffman, Bethlehem Steel Co., Sparrows Point, compared a 28 ft. furnace with a 19 ft. furnace. Both were run with the same raw materials for one month. Both furnaces produced record tonnages during the month in question, the 28 ft. furnace, 1753 tons per day; the 19 ft. furnace, 893 tons per day.

The standard deviation of the silicon in the iron was larger on the small furnace. Eighty-six per cent of the casts were within the most desirable silicon range on the large furnace, only 63 per cent were on the small furnace. The large furnace showed a coke rate 14 lbs. per ton of iron lower than the small furnace. Mr. Hoffman shortly touched the controversial point of rating furnaces and expressed the opinion that the productivity of the large furnace was greater. He also found tapping hole lengths and maintenance equal on both the large and the small furnaces.

Mr. Plimpton of Inland Steel Co., East Chicago, discussed rating of furnaces. He described the old method of rating furnaces on the basis of square feet of hearth area. This was followed by Rice's method of figuring on the basis of a 6 ft. annular ring before the tuyeres, and by Hoffman's method using a 3½ ft. annular ring. The 6 ft. ring favors the small furnaces while a 3½ ft. ring proved to be almost directly proportional to the tonnages in good and in poor operating periods at Inland. Mr. Plimpton referred to 8 furnaces 17 ft. 3 in. to 25 ft. 9 in. in hearth diameter.

Mr. Plimpton showed figures indicating better coke rates at the larger furnaces, these also showing more consistent quality and lower flue dust production. The tapping holes of the large furnaces stood up at least as well as the ones of the small furnaces.

Mr. Plimpton pointed out that large furnaces appear to get into a one-sided condition of working much more frequently than small furnaces do. This is shown by hot cinder from one monkey and cold cinder from the monkey on the opposite side of the furnace. It is also shown by a greater spread of inwall temperatures. Because of the greater area at all levels of a large furnace it is easier for a maldistribution of materials to take place with a resulting maldistribution of wind. This, however, has proved to be not particularly serious since a lopsided condition can be corrected by filling changes.

Mr. Plimpton presented data substantiating Mr. Burks' contention that large furnaces suffer more from inferior raw material than small furnaces do. Sized anthracite was used on a 25 ft. 9 in. furnace replacing 5 per cent of the furnace coke, and then for a few weeks replacing 10 per cent of the furnace coke. With 5 per cent anthracite, the wind had to be reduced 3.2 per cent and the average daily tonnage was reduced 38 tons; with 10 per cent anthracite there was a further reduction in wind of 3.5 per cent and a further reduction of tonnage of 28 tons. At the same time there was an increase of copper losses and in tap hole trouble due to excessive amounts of kish, the same symptoms that occur when the strength of coke drops. This was to be expected, since the standard tumbler test of the anthracite coal used showed a stability index of only 6 per cent compared with 45 per cent for the furnace coke. Later 5 per cent and 10 per cent anthracite was used on two small furnaces of 17 ft. 3 in. and 19 ft. hearth diameter. There was no loss of wind or tonnage in either case.

It might be noted that, except for the question of uniformity of pig iron, the speakers were in agreement.

The Proposal to Increase Dues

To ask a member of any organization whether or not he wishes to have his dues raised invites the obvious answer. It is the same answer he would be likely to give if her were asked if he wished to have his taxes raised. However, in each case, if the increased money thus supplied were used wisely, the person might personally benefit by considerably more than the amount of the increased levy; in which case his answer should be in the affirmative.

The American public has pretty well accepted practically a doubling in the price of many things in the last few years, food being a conspicuous and widely purchased example, automobiles another. A few things, such as rent, have increased comparatively little, being under government control, but those who furnish these services generally have been able to get by through continuing peak use of the facilities they provide. Combining all the important things that one has to buy, the Government tells us that the index of the cost of living last year was about 170, based on 100 for 1935-39, which means the purchasing value of the dollar dropped to 59 cents.

Everybody knows all about this. But the fact remains that it is only human to wish to hold one's expenses in check wherever possible, and conversely to increase one's income by such means as are available. Everyone would like to have the difference between his income and his absolutely necessary expenses as large as possible, so that he can have the maximum amount to do with as he likes, be it buying a few books, a television set, going to the theatre oftener, patronizing better restaurants, or maybe improving his brand of liquor.

Taxes one must pay, willy-nilly. Dues to the AIME one need not pay. An individual may drop his membership and pay nothing. As the majority wills, members may continue to pay the same amount as at present, or pay a $2 to $5 annual increase for three years. The return, in the latter case, however, will be considerably more than proportional.

To illustrate specifically how a service can be provided all members for a nominal cost that individually would be exorbitant, data concerning the distribution of the 1947 AIME Directory will help. Aside from the necessary cost of keeping address files up to date, a Directory may be published and distributed to all members for a little less than a dollar each. Experimenting with the idea that certain Institute services should be provided only for those who want them at additional cost, the 1947 Directory was sent only to members who were willing to pay one dollar for it. Twenty-three hundred paid for copies of the Directory. Due to the apparent limited demand for a Directory every year even at the nominal charge of one dollar, and to the need for conserving cash, no Directory will be published in 1949. On the basis of 2300 orders the 1947 Directories cost the Institute about $4.50 each (5000 were printed). At $4.50, or even at $2 to include over 2000 distributed free to new members, committee chairmen, and secretaries, fewer still would buy the Directory and the cost per copy would be exorbitant. Apparently, when issued, Directories should go to all members as part of their dues and it simply remains to decide how often the members as a whole wish to have part of their dues used for a Directory.

This shows how much more economically services can be provided to all members than to comparatively small groups. Less obviously, perhaps, it illustrates that when the purchasing value of the income from dues decreases, tangible services have to be cut excessively.

If a man who is active professionally invests from $25 to $50 per year annually in a membership in the AIME, a subscription to one or two other technical journals in his field, and a few of the new books on his specialties, he should get ideas, knowledge and friends worth hundreds or even thousands of dollars to him.

All of which leads up to the current question before AIME members, and on which most of them will soon be asked to vote: Shall the dues of Members and Associate Members be raised five dollars and of Junior Members two dollars for the years 1950, 1951, and 1952? A statement to the entire membership appears on the first page —page 181—of the All-Institute section in this issue.

Important Actions Taken by Executive Committees *(See page 181, Section 2)*

Metals Branch Council Formed

One of the most important actions taken by the Executive Committees of the Metals Divisions at the Annual Meeting was the confirmation of the plans for organizing the Metals Branch Council. Now that the Institute is to be completely divisionalized and the Divisions broadly grouped into three Branches some organization at the Branch level was necessary. There is no intention of disturbing the semiautonomy of the Divisions but a means of coordinating the activities of similar Divisions seemed desirable. With the formation of the Extractive Metallurgy Division to include essentially nonferrous extractive metallurgy and in particular those committees which dealt with the reduction and refining of nonferrous metals, the Metals Branch became completely divisionalized. In addition to the Extractive Metallurgy Division, the Metals Branch includes the Iron and Steel Division dealing with the production of iron and steel and the Institute of Metals Division which now includes all physical metallurgy—both ferrous and nonferrous metallurgy of the solid state.

The Metals Branch Council was established in San Francisco as follows: The Chairman and past Chairman of each of the three metals divisions automatically represent these divisions. One representative each will be appointed to the Metals Branch Council by the Mineral Industry Education Division and the Mineral Economics Division to represent these Divisions which are Institute wide and not confined to any one Branch. The Chairman of the Metals Branch Council is to rotate among the past chairmen of the three metals divisions starting with the oldest division. The AIME assistant secretary in the metals field is to act as secretary according to current plans. It is to be noted that since Division Chairmen are AIME Directors, the Metals Branch Council is closely coordinated not only with the Divisions but with the AIME Board of Directors also.

Probably the most important activity of the Metals Branch Council will be its responsibility for the JOURNAL OF METALS, the monthly journal of the Metals Branch and its three Divisions. The Transactions section of the journal is now closely supervised by the Technical Publications Committees of the Divisions. An Editorial Advisory Committee for sections 1 and 2 has just been appointed. The Metals Division Chairmen serve ex-officio with two other representatives of each of the three Divisions. Note is taken here of the AIME Board of Directors authorization to hire a full time advertising manager for the JOURNAL OF METALS. Every effort is being made to secure a sufficient amount of advertising in the JOURNAL OF METALS to bear a large part of its cost as is now the case with Mining Engineering.

Several weeks ago, those who had selected the JOURNAL OF METALS and others who were believed might have an interest in one of the metals Divisions were mailed business reply postcards so that we might have a record of the membership of the new Extractive Metallurgy Division as well as a record of the changes resulting from the revised scope of the IMD and ISD. Of the 6000 cards mailed 3800 have been returned.

MAJOR INTEREST
(One Selection Only)

	Actual Returns	Estimate for 100 % Return
IMD	1316	2072
ISD	880	1387
EMD	906	1428

DIVISIONAL AFFILIATION
(Multiple Choice Permitted)

	Actual Returns	Estimate for 100 % Return
IMD	1928	3035
ISD	1480	2332
EMD	1358	2140

IMD Executive Committee Stresses Value of Dues Increase

In order to provide more time and a larger participation in the discussion of the current basic policy decisions both the IMD and ISD Executive Committees scheduled special meetings following their regular Annual Meetings. At their March 22nd meeting the IMD Executive Committee noted the large number of technical papers being published by the AIME in its field and went on record as unanimously favoring the three year increase in dues to make possible publication of these excellent papers in the current period during which advertising revenue is being built up to bear a portion of this expense.

Plans were made to help the new advertising manager as much as possible.

Authors are encouraged to enhance their reputations by carefully writing concise papers. The task of keeping up with the large volume of metallurgical literature is thereby made easier for the readers and your Institute is less burdened by unnecessary expense. Extensive reviews of the literature need not be given at the beginning of papers if reviews have recently appeared elsewhere. Supposition and the asserting of new theories are frequently of considerable value in developing the science. However, new theories can be presented briefly and detailed discussion withheld until sufficient data is accumulated to justify detailed consideration.

A recommendation was made to the program committee to organize forums on the science of metals at the Annual Meeting. These forums would be off-the-record and papers would not be preprinted. Opportunity would thus be provided for presenting theories and exchanging ideas on which any one experimenter had insufficient data to be conclusive. Such an exchange might help an experimenter avoid repeating experiments found fruitless by others as well as guide him in promising directions.

Howard Bruce to Speak at ISD May Dinner

Howard Bruce, Deputy Administrator, Economic Cooperation Administration, will speak at the Annual Iron and Steel Div. May Dinner 7 p.m. May 25 on "ECA and You" in the Georgian Room, Statler Hotel, New York. Cocktails will be available at 6 in the foyer of the Georgian Room.

Besides Iron and Steel Division members, all AIME members and guests are welcome, at $6.00 per person. Return post cards were sent to ISD members east of the Mississippi, but if you did not get a card for this stag dinner, write the secretary of the Iron and Steel Division at American Institute of Mining & Metallurgical Engineers, 29 W. 39th St., New York 18, indicating the number of reservations desired.

Iron and Steel Division Executive Committee

Actions taken by the Iron and Steel Division Executive Committee at its special meeting April 20th were not available when this issue went to press. At the Annual Meeting the Reduction of Ferroalloy Ores Committee, previously a nondivisional Committee, was officially made a Committee of the ISD. Both the similarity in reduction methods to that of iron ore and the fact that ferroalloys are primarily used in iron and steel production indicated this to be a logical course. The ISD Committee will simply be known as the Ferroalloy Committee. It is intended that the activity of this committee be increased and its scope broadened to include uses of ferroalloys as well as the reduction of their ores.

Chairman King pointed out that Committees of the Coal Division, the ISD and the IMD pretty well look after securing papers from the raw materials to finished product except for hot rolling steel. A committee was authorized to study the possibilities for a committee on hot forming steel to stimulate papers and experimental work in this field of the ISD.

Although not peculiar to the ISD, the Executive Committee discussed the present procedure whereby a Student Associate does not transfer to Junior Membership for a year or more after he leaves school. As a result many lose interest and contact in the interval and drop out for several years to their and the Institute's disadvantage. A recommendation was made to the AIME Admissions Committee that a procedure be developed whereby Student Associates are encouraged to request a change of status to Junior Membership a month or so before graduation.

Cleveland AIME Hears Aluminum Hardening Talk

A large attendance was present at the Cleveland Section A.I.M.E. March meeting to hear about the "Practice Aspects of Precipitation Hardening in Aluminum Alloys." The talk was ably presented by Harold Hunsicker, a research metallurgist at the Aluminum Co. of America.

The history of precipitation hardening was reviewed, starting with the discovery of Wilm in Germany. The general requirements for this type of hardening were discussed, and the changes of yield strength, tensile strength, hardness, lattice constant, resistivity and volume change with time and temperature of aging after solution treatment were strikingly illustrated with colored lantern slides. The change of solution potential after aging is of considerable importance in the corrosion behavior of precipitation hardened aluminum alloys.

Mr. Hunsicker showed that when the growth characteristics, residual stresses from quenching, hardness, and other properties are known as functions of time and aging temperature, a heat treatment for optimum resulting properties can be worked out. Examples of precipitation hardening applied to "as cast" materials without the need of a separate solution treatment were given. Precipitation hardening was shown to be an extremely complex phenomenon involving the integration of a series of factors, some tending to harden and some to soften the alloy.

An interesting "inverse" application of precipitation hardening was given, where hardness measurements of an aircraft piston after use indicated the temperature reached in service when the time — temperature — age hardening characteristics of the alloy are known.

VERNON F. KALAN

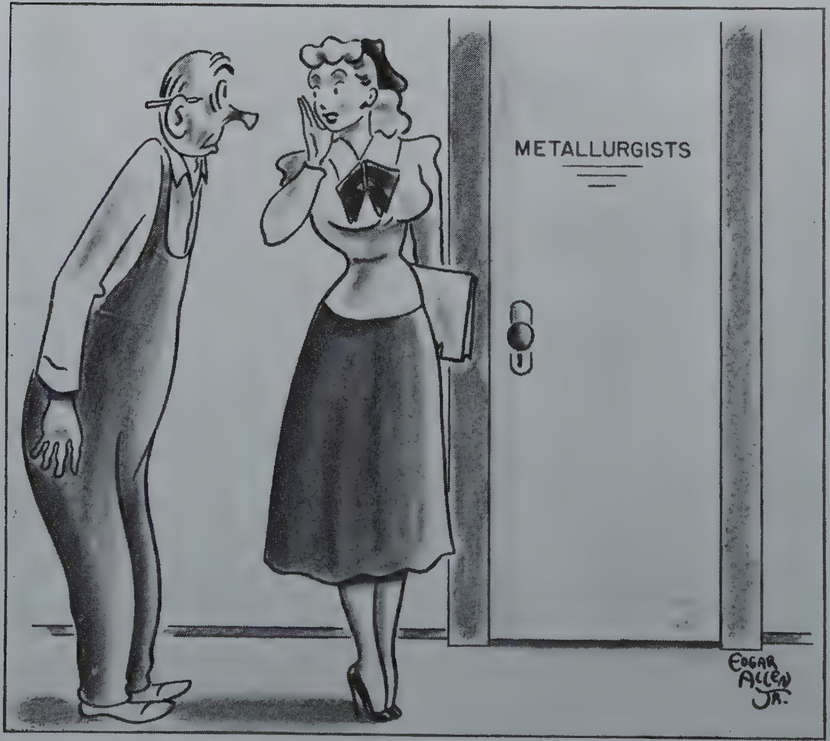
Detroit AIME Hears Talk on Electric Ingot Process

The Kellogg Process of Ingot Production was discussed by R. K. Hopkins, Manager, Electric Products Dept., M. W. Kellogg Co., before the Detroit Section, AIME, on March 21. Frederick P. Bens, Climax Molybdenum Co., was chairman of the session.

The Kellogg process is an outgrowth of a continuous melting and cladding method developed for the production of composite corrosion-resistant materials to handle the "sour" crudes of the petroleum industry. The alloy-making portion of the cladding method was separated from the rest of the process and adapted to the melting and casting of alloy ingots that are literally pipeless. Mr. Hopkins presented the physical properties and the results of metallographic examination of 16-25-6 alloy produced by this process to illustrate the uniformity and quality of the product. While this specialized method cannot compete with normal low-carbon grades of SAE steels, it is extremely

SMELTS and SMILES

By Edgar Allen, Jr.



"Should I go in? Is metallurgists a contagious disease?"

valuable for the production of high-alloy steels for applications under severe service conditions of corrosion or high temperature.

For example, in spin tests of turbine wheels so designed that failure would result in blowing up of the wheel, materials made by conventional metallurgical practice failed at 32,000 r.p.m., whereas wheels made by the electric ingot practice failed at 40,000 r.p.m. when the wheels went out of balance. Mr. Hopkins associated the improved ductility of the material made by the new process to the minimum porosity and segregation of the product. In this connection he pointed out that the critical point of the process was the development of a suitable slag. Although used predominantly for high-alloy steels, this process has been used for melting nonferrous compositions.

In the discussion that followed, the members of the Detroit Chapter expressed great interest in two developments which resulted from the electric ingot process. The first was the electric hot-top process by which it is claimed that steel-making capacity can be increased without increasing ingot size through practically complete elimination of pipe. The second process was a method of producing a transition piece for the assembly of power plant steam lines.

In conclusion Mr. Hopkins mentioned that seamless tubing blanks have been made by casting around a water cooled copper core. The advantages of this process are not yet sufficient to warrant substitution for the present commercial methods.

M. SEMCHYSHEN

The total tonnage of acid open hearth steel melted in 1948 for both castings and ingots was 1,267,177 tons, according to the Acid Open Hearth Research Assn., Pittsburgh. Of this total 49.15% represents ingots and 50.85% castings.

Tonnages for five years:

1944	2,107,493
1945	1,542,724
1946	1,132,199
1947	1,257,673
1948	1,267,177

(From the secretary, Fred C. T. Daniels, Mackintosh — Hemphill Co., 901 Bingham St., Pittsburgh 3).

BLAST FURNACE COPPER CASTINGS

Tuyeres - Tuyere Coolers
Cinder Notches - Notch Coolers
Bosh Plates - Mantle Plates
Hot Blast Valves and Valve Seats

Standard, Special and most Patented Designs to your specifications.

SMEETH-HARWOOD COMPANY
2401-09 West Cermak Road, Chicago 8, Illinois

Superior Blast Furnace Copper Castings Exclusively

Heavy Nose, Super-Cooled Bosh Plates—for the Tuyere Zone where plates are most apt to burn.
Cinder Notch
SMEETH-HARWOOD Hot Blast Valve Seats — the recognized standard of dependability
Refractory Lined Safety Coolers
Steudel Tuyere shaped to eliminate cutting underside of nose

"C-RO" POLISHING COMPOUND

"C-RO" is a relatively heavy, metallographic, polishing compound in which small, sharp uniform particles provide a fast, sure and flawless cutting action. It polishes without burnishing. Write for Bulletin 208.

53-585 . . . . "C-RO" . . . . ¼ lb. $5.00

BURRELL
1942 FIFTH AVE
PITTSBURGH 19
PENNSYLVANIA

INDUSTRO
model B
GAS ANALYZER

Analyzers for the rapid control analysis of gas mixtures comprising any or all of the following: carbon monoxide, methane, oxygen, hydrogen, and nitrogen. Two models, B and C, for use in combustion control plus other applications. Write for Bulletin 213.

Model B		**$87.50**
Model C	**110 volts**	**$140.00**

BURRELL
1942 FIFTH AVE
PITTSBURGH 19
PENNSYLVANIA

Engineering Societies Personnel Service, Inc.

New York—8 West 40th St., Zone 18.
Detroit—100 Farnsworth Ave.
San Francisco—57 Post St.
Chicago—84 East Randolph St., Zone 1

These items are from information furnished by the Engineering Societies Personnel Service, Inc., which is under the joint management of the four Founder Societies. This Service is available to members and is operated on a co-operative, nonprofit basis.

In applying for positions advertised by the Service, the applicant agrees, if actually placed in a position through the Service as a result of these advertisements, to pay a placement fee in accordance with the rates as listed by the Service.

When making application for a position include six cents in stamps for forwarding application to the employer and for returning when necessary.

All replies should be addressed to the key numbers indicated and mailed to the New York Office.

A weekly bulletin of engineering positions open is available to members of the cooperating societies at a subscription of $3.50 per quarter or $12 per annum, payable in advance.

POSITIONS OPEN

PLANT MANAGER, 40-50, graduate chemical or metallurgical engineer, to take complete charge of production for the separation and refining of non-ferrous metal ores, including the roasting operations and recovery of other elements contained therein. Salary, $8,000-$10,000 a year, plus bonus. Location, northern New Jersey. Y1871.

METALLURGIST, under 40, with smelting and refining background for research work. Should have worked in non-ferrous metals plant. Salary, about $4,800 a year depending on experience. Location, Texas. Y2087.

RESEARCH FELLOWSHIP in Metallurgy at an eastern university. Post graduate work for M.S. or Ph.D. degree. Selection of research problem in variety of fields including welding, induction heating, corrosion, powder metallurgy and electrometallurgy. Starts September 1949. Salary, $1,200 a year if single; or $1,800 a year if married, plus tuition fees. Location, Pennsylvania. Y2089.

METALLURGIST, 25-35, with college training, including and preferably with emphasis on steel melting practice and with two years' or more actual practical experience in steel melting. Duties will be to make sure that correct melting procedure as laid down are carried out and to supervise work of observers in both open hearth and electric melting departments. Salary about $4,200 a year. Location, Pennsylvania. Y2245.

RESEARCH METALLURGIST, preferably with some experience in physical metallurgy for powder metallurgy work. Salary, $3,000-$3,600 a year. Location, New York Metropolitan Area. Y2254.

METALLURGIST OR MINING GEOLOGIST with broad experience in the economics of metals and minerals, to contact industrial executives and government officials, and present his findings in well-written reports. Salary, $4,500-$6,000 a year depending on qualifications. Y2264D(a).

METALLURGICAL ENGINEERS, (Req. 588), 26-35, with B.S., M. S. preferred, in metallurgical engineering. If majored in metallurgy, degree in chemical engineering acceptable. Should have five to ten years' experience in metallurgical research. Experience in powder metallurgy preferred. Experience in high-vacuum techniques helpful. Position involves development of pure metals and research work in the field of powder metallurgy, preparation of work outlines and research reports. Salary, $4,200-$5,040 a year. Location, New York Metropolitan Area. Y2273.

TEACHING PERSONNEL. (a) Instructor or Assistant Professor with some teaching and/or industrial research experience, to teach ferrous physical metallurgy. Excellent opportunity for research. (b) Instructor or Assistant Professor, experienced in metallurgical X-ray diffraction work and qualified to teach that subject at the graduate level. Salaries open dependent upon qualifications. Location, Indiana. Y2292.

MEN AVAILABLE

METALLURGICAL ENGINEER (Met. E. degree Columbia), 25 years' broad experience; ferrous and non-ferrous foundry management; quality control problems; development of non-ferrous alloys; consulting experience. Executive ability. Desires position as foundry manager or director of metal development or research. M-428.

Advertisers' Index

	Page
Bell Telephone Laboratories, Inc. *N. W. Ayer & Son, Inc.*	Inside Front Cover
Buehler Ltd. *Kreicker & Meloan, Inc.*	4
Burrell Technical Supply Co.	19
Federated Metals Div., A.S. & R. Co. *John Mather Lupton Co., Inc.*	Inside Back Cover
Fisher Scientific Co. *Tech-Ad Agency*	Outside Back Cover
International Nickel Co., Inc. *Marschalk & Pratt Co.*	2
Molybdenum Corp. of America *Smith, Taylor & Jenkins, Inc.*	6
Morgan Construction Co. *The Davis Press, Inc.*	1
National Research Corp. *Henry A. Loudon Advertising, Inc.*	5
Smeeth-Harwood Co. *Allbright Associates Advertising*	19

PROFESSIONAL SERVICES

Limited to A.I.M.E. members, or to companies that have at least one A.I.M.E. member on their staffs. Rates, hereafter, one inch only, for one year, $40.

R. S. DEAN LABORATORIES
Consulting, Research, Development
Chemistry, Electrochemistry, & Metallurgy
Laboratory Research on a Contract Basis
2005 K St., N. W. EX 5656
Washington 6, D. C.

MAX STERN
Consulting Engineer
Expert for Scrap Recovery and Shipwrecking—Modernization of Plants and Yards for Ferrous and Nonferrous Metal Scrap
150 Broadway New York 7, N. Y.

C. L. MANTELL
Consulting Engineer
Tin Metallurgy
Electrochemical Processes
451 Washington St., New York 13, N. Y.

LEWIS B. LINDEMUTH
Consulting Engineer
140 CEDAR STREET, NEW YORK 6, N. Y.
STEEL PRODUCTION
DESIGN · OPERATION · CONSTRUCTION · METALLURGY

ALL INSTITUTE SECTION

APPEARING IN:
- MINING ENGINEERING
- JOURNAL OF METALS
- JOURNAL OF PETROLEUM TECHNOLOGY

To Members of the AIME:

For three years your Institute has been operating at a deficit, caused primarily by the increased cost of printing. This deficit has been largely met by borrowing from certain of the Funds which have been established for specific purposes by their contributors. Including an estimated $50,000 loss in operations for 1949, these borrowings will have reached the staggering sum of $150,000 by the end of the current year. Obviously, the Institute cannot continue to function on this basis.

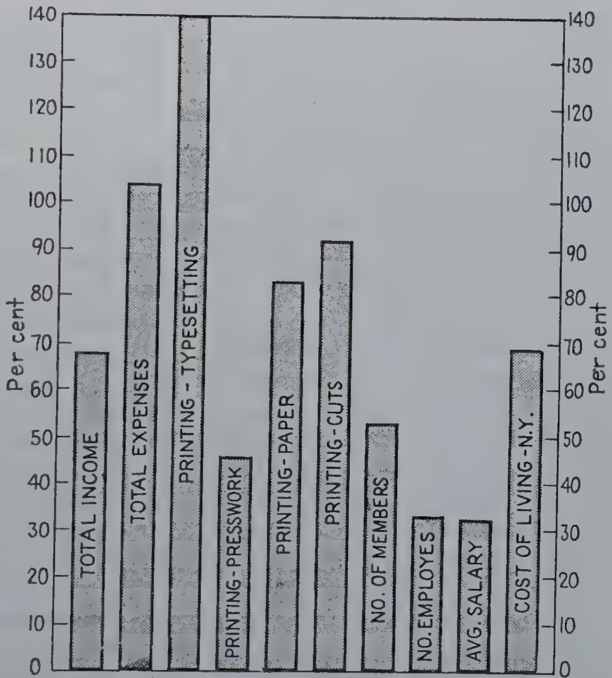
The Institute's main sources of revenue are advertising and membership fees. Every effort is being made to increase advertising revenue and to this end a new publication policy has been adopted whereby we put out three magazines, each in specialized fields, instead of *Mining and Metallurgy* and the four *Technologys* as heretofore. When these magazines become well established we have every reason to expect that the advertising revenue will be considerably greater than it was, but it may take three years to reach this goal. As to membership dues, a comparison with those of the other Founder Engineering Societies shows that ours are the lowest.

Taking due account of these two factors, the Section delegates at the San Francisco meeting recommended to the Board of Directors that a ballot be submitted to the membership to provide for an increase in dues, for a three-year period commencing in 1950, of $5 for Members and Associate Members and $2 for Junior Members. The Board of Directors unanimously approved the recommendation of the Section delegates.

It is estimated that this increase in dues will amount to approximately $50,000 per annum, which would permit a balanced budget in 1950 based on 1949 estimated income and expenditures. At the end of the three-year period another referendum vote will be taken, at which time dues for the succeeding years will be determined, based upon the conditions then prevailing.

A ballot will be sent to each voting member of the Institute in the latter part of May and you are urged to vote affirmatively in accordance with the recommendation of the Section delegates and the Board of Directors.

L. E. YOUNG
President



Percentage increases of various items 1948 over 1939.

Big Turnout of Directors Studies Vital Institute Problems at Six-Hour March Board Meeting

Nineteen Directors—a majority of the Board—spent six hours with their sleeves rolled up discussing important Institute problems at the meeting in New York on March 16. One out-of-town Director wrote back after returning home that he would come to the next meeting prepared with a knapsack and sleeping bag. A new procedure was tried for the first time: the usual monthly luncheon meeting of the Executive and Finance Committees was dispensed with, and instead the entire Board gathered at the Engineers' Club for a 12:30 luncheon and informal session without guests. Then two hours later, the regular open meeting got under way, and lasted until ten minutes to seven. President Young presided, the other Directors present being Messrs. Alford, Daveler, Elkins, Head, King, Kinzel, Kraft, McLaughlin, Meyerhoff, Pehrson, Peirce, Phillips, Price, Rhines, Schumacher, Suman, Swift, and Fletcher. T. B. Counselman acted as alternate for S. J. Swainson. In addition, absentee comment on various items on the agenda was received from Oliver Bowles, H. N. Eavenson, Louis D. Gordon, and T. S. Washburn.

more advertising income

One of the vital matters discussed at length was the problem of increasing the income of the Institute through advertising in the new journals. Considerable opposition had developed to the suggestion that an over-all advertising manager, or publication manager, be secured, at least for the present. The Petroleum Branch, especially, wished to contiuue its present efforts to build up advertising with only its staff at the Dallas office. The problem was believed most critical with the *Journal of Metals*, and authority was voted to secure an advertising solicitor immediately for this magazine. Mr. Schumacher, chairman of a small committee that had been appointed to consult with advertising men, reported their views, and Theodore Marvin gave the group the benefit of his experience as both a "space" salesman and buyer. Wheeler Spackman reported increased billings in the first quarter for *Mining Engineering* compared with *Mining and Metallurgy* last year, in spite of the loss of some advertising to the *Journal of Metals*. Advertising income for the three new journals should, it is believed, exceed that of last year by at least $30,000. "Publishers' representatives" may be used to augment regular staff coverage.

proposed dues increase

Another matter of great importance was consideration of the action of the Section Delegates at the San Francisco meeting, recommending to the Board an increase in the dues of Members and Associate Members of $5, and of Junior Members of $2, for three years ending with 1952. The Committee on Ways and Means had been asked to study this proposal, and Mr. Kraft, chairman, presented a summary of the financial condition of the Institute. If advertising income could be raised to $85,000 this year, and certain economies made, the deficit might be reduced to about half what it was in 1948, or some $42,000. This would bring the accumulated deficit for recent years up to about $150,000. The proposed increase in dues should increase yearly income by $55,000 to $60,000 for the three years, which would slightly more than offset the borrowings that have been necessary to meet the deficit. In 1952 a referendum would be taken as to whether or not the increased dues should be continued in the light of conditions at that time. The Board voted to accept the recommendation, and to conduct the required referendum of the membership. If the vote favored the increase, the necessary change in the bylaws would be voted on at the June meeting of the Board.

Other proposed changes in the bylaws to be acted upon by the Board at its June meeting will restore Jan. 1 as the date when dues become payable, with three months of grace allowed before members cease to be in good standing and the journals are cut off; and will dispense with the grade of Junior Foreign Affiliate at the end of 1949. The date when dues become payable had been changed to Oct. 1, but to increase the efficiency of work in the Institute's accounting department it is felt that mailing of the bills should be staggered through the months of September, October, November, and December.

Annual Meeting in St. Louis?

Shortly before the Board met, an invitation had been received from the St. Louis Section, the St. Louis Chamber of Commerce, and the St. Louis Convention and Publicity Bureau, inviting consideration of St. Louis as the scene of the Annual Meeting in 1951. Time had not permitted mature consideration of the proposal, so acceptance was postponed until the April meeting of the Board.

Mr. Schumacher, chairman of the Committee on Local Section Appropriations, reported his difficulties in making the available money—which the Board had set at approximately the same figure as last year, $9000—meet the needs of the Local Sections. Many had offered excellent arguments for increased allocations, up to $600 each though the Bylaws set a limit of $400. The Board was sympathetic to the needs of the Local Sections, many of which were believed to be doing an exceptionally fine job, but, with the depleted finances of the Institute, could not see its way to raise the total appropriation. The Local Section reports were studied carefully, and the fairest division possible was made. However, to alleviate the situation to some extent for the Local Sections, each was voted 10% of the initiation fees paid by all new members in its territory. This will be effective with the April elections. Thus a Local Section will receive an extra $100 if fifty new Members or Associate Members are elected from its area in the course of a year.

cost-cutting ideas

Mr. Elkins reported that the Petroleum Branch was planning issuance of a Directory of its members in the early summer, to be financed from its own funds. No complete Institute Directory is to be published until 1950. He also offered various recommendations designed for increased economies and efficiencies in the conduct of Institute affairs, suggesting that committees be appointed to study these matters. He was assured, however, that the Board and staff had already given close study to some of the suggestions and were keenly alive to the need for cutting costs in every way possible consistent with maintaining the essential services of the Institute. Particularly, Mr. Elkins thought that all meetings should be self-supporting in justice to those who could not attend them. In this the Board fully concurred, and voted that registration fees and other charges at all Annual, Regional, and Divisional meetings be set at such figures that they would be self-supporting to the fullest extent possible.

Chairmen of the Institute's standing committees for 1949-50 were announced and the complete committee personnel, with Local Section and Student Society officers, will be published in the June issues of the three journals. Certain inactive committees will be discontinued.

The Committee on Nominations for Institute Officers reported some difficulties in interpreting the rules and selecting candidates that properly represented the professional and geographical divisions of the membership. President Young was therefore authorized to appoint a committee of three to revise and clarify the "Instructions and Advice" to members of this Committee.

segregating Transactions material

Some members had reported a preference that the Transactions volumes be published by Divisional subjects, as in the past, rather than combining all the technical publications of a Branch, as is the plan for the future. Mr. Price, especially, reported the wish of the Coal Division that its papers be segregated, both in *Mining Engineering* and in a Transactions volume. This could be done, it was reported, either by holding the type for a year or more, or by separate offset printing of the desired papers, but either would be more costly than the present practice. The Board felt that the matter should be reconsidered before the end of the year, with relative costs considered.

Suggested plans for special committees of the Directors to consult frequently with the editorial boards of the three Branch journals were discussed, but it was felt that if Directors keep in close touch with the officials of the Branch they represent, and the Secretary keeps himself apprised of the wishes of members at the Local Section level, no further organization should be necessary. A continuous sampling of reader opinion on *Mining Engineering* is under way for the guidance of the editorial and advertising staffs.

delegates and representatives

Among the routine matters disposed of were the following: W. H. Carson was appointed an official delegate to the World Engineering Congress in Cairo. E. H. Robie was appointed as a director of the Engineering Societies Personnel Service. A. Rodger Denison was appointed an AIME director of the American Geological Institute for the term ending November, 1949, and Joseph L. Gillson for the term ending November, 1950. Approval was given the proposal to invite G. Borelius, a professor at the Techniska Högskolan, Stockholm, as Institute of Metals Division Lecturer for 1951. Francis B. Foley was appointed to represent the AIME on the Engineers Joint Council Committee on Collective Bargaining by Engineers in Professional Work.

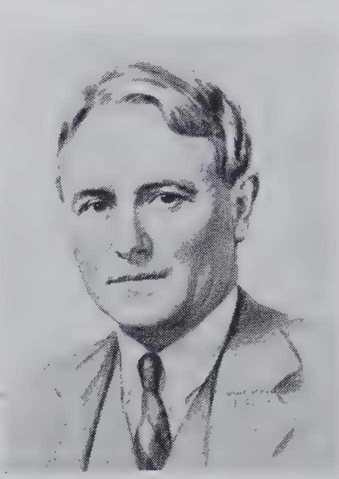
Howard A. Herder was named to represent the AIME on the Washington Award Commission for a two-year term, to May, 1951. Gail F. Moulton was appointed to succeed Ralph M. Roosevelt as an AIME representative on United Engineering Trustees for a four-year term expiring in October, 1953. R. M. Brick and Francis B. Foley, with A. W. Grosvenor as alternate, were appointed official delegates to the meeting of the American Academy of Political and Social Science at Philadelphia, April 8-9. Approval was given to the continuance of the Alloys of Iron Research as an AIME-sponsored Engineering Foundation project for 1950 and 1951. A price of $3 to members and $5 to nonmembers was set for the translation, in one volume, of the old German books, "Probierbuchlein" and "Bergbuchlein," soon to be published.

L. E. Young was named to represent the AIME on the John Fritz Medal Board of Award for a four-year term ending in October, 1953. Attention was called to the fact that it is the turn of the AIME representatives to name a candidate for the Fritz Medal this year, and nominations are invited. Suggestions may be sent to Harvey S. Mudd, Louis S. Cates, Erle V. Daveler, or W. E. Wrather.

The sincere thanks and appreciation of the Board was voted to the San Francisco Section for its conduct of the Annual Meeting, and thirty copies of the 75th Anniversary Volume of the AIME were ordered sent to William Wallace Mein, Jr., Chairman for distribution to committee chairmen and other active workers.

Donald H. McLaughlin Nominated for President in 1950

Donald H. McLaughlin, president of the Homestake Mining Co. and a Vice-President of the AIME, has been named to head the ticket for new Board members in 1950 by the Committee on Nominations for Institute Officers. Nominations for Vice-President are Andrew Fletcher and Robert W. Thomas, both of whom currently hold the same office, but whose present terms expire next February. The six other Directors chosen are: Francis B. Foley, superintendent of research for The Midvale Co., Philadelphia; Edward Meagher, treasurer of the Texas Gulf Sulphur Co., New York; Howard I. Young, president of the American Zinc, Lead and Smelting Co., St. Louis; Harold Decker, assistant manager of the Pan American Production, Pipe Line, and Gas Co., Houston; and C. V. Millikan, Amerada Petroleum Corp., Tulsa. Of these, only Mr. Millikan is currently a member of the Board.



Donald H. McLaughlin

At its June meeting the Board will name a ninth candidate, and if no supplementary tickets are named by Sept. 1 the above slate is automatically elected. In that case, a new Vice-President will also have to be elected by the Board, owing to the elevation of Dr. McLaughlin to the Presidency.

Where to Split the Metallurgists?

Is an operator of a cyanide plant, a copper leaching plant, a roasting plant, or a mercury distillation plant primarily a mineral beneficiator or an extractive metallurgist? The officials of the Minerals Beneficiation Division and of the new Extractive Metallurgy Division each claim these and some other similar fields, in whole or in part. The practical difficulty, so far as the AIME is concerned, is that the technical papers of the former Division are to be published in MINING ENGINEERING and of the latter in the *Journal of Metals.* Even some of the individual members immediately concerned are by no means certain as to whether they are more closely allied to the Mining or to the Metals Branches. into which the Institute is now broadly divided. A pyrometallurgist, working in a copper smelter, a field of activity not even claimed by the MBD. writes that he works in a mining district, all his friends are mining and millmen, he cannot understand and has no interest in the type of papers published in *Metals Technology,* and would like to see all smelting papers published in *Mining Engineering.* But most members of the Committee on Reduction and Refining of Copper do not appear to agree with him.

Much thought has been given to the problem but no thoroughly satisfactory solution has been found. Authors' wishes as to the medium of publication will be considered, but ultimate decision will rest with the Technical Publications Committee. For the time being, members in the controversial fields should pick the journal that they think likely to be of major interest to them, and then watch the list of all technical papers published. If they note one in which they are interested published in a journal to which they do not subscribe, they can obtain a copy for 75 cents. Fortunately it does not seem likely that cases of this kind will be too numerous.

Expensive Advertising

A double-page spread in the February *Fortune* is devoted to the wonderful results that advertisers get in that magazine. Among other items is: "A page in the August issue resulted within ten days in a closed deal for 3,000 square feet of flooring." Business must be pretty bad.

New Arizona Scholarships

A grand total of twelve scholarships at the University of Arizona and eight at Arizona State College have been established for students of mining, metallurgy, and allied fields by the Phelps Dodge Corp. One first year scholarship will be given each year at the State College, and during the first year, comparable grants will be set up for the second, third, and fourth years. The plan at the University of Arizona is the same, but double the number of awards will be made, and graduate studies are also included. Eligibility is limited to Arizona residents.

Paul G. Hoffman to Speak at ISD May Dinner

Paul G. Hoffman, ECA administrator, will be the speaker at the annual Iron and Steel Division May dinner on May 25, his subject, The ECA and You. The dinner will be held at the Statler Hotel in the Georgian Room at 7 p.m., preceded by a social hour in the foyer to the Georgian Room. Tickets are $6. Please make reservations through the Secretary of the ISD, AIME, 29 West 39th St., New York City 18.

Correction, Please

We received the following letter from R. G. Knickerbocker, chief, Rolla branch, metallurgical division, Bureau of Mines. We wish we could accept Mr. Knickerbocker's kind explanation of the error, but fear that our printer might get angry and any way we were brought up to be relatively honest.

"In the interest of the manganese industry, and especially the Electro Manganese Corp. at Knoxville, Tenn., I believe it will be necessary to correct the printer's mistake made in the fourth paragraph of the second column on page 97 under electrolytic manganese. (March Annual Review on Ferroalloy Metals.) The price of electrolytic manganese effective Jan. 1, 1949, has been reduced to $0.28 per lb and not to $1.28 per lb. No doubt this is perfectly clear to those technically concerned with the manganese industry but to a great many others concerned with the critical status of treatment of low-grade domestic manganese ores this figure of $1.28 might be extremely misleading."

1949 Annual Banquet Souvenirs

The miniature gold pan, made of sterling silver, commemorating the discovery of gold in California, has caught the eye of many a member.

The pins were presented to all ladies attending the 1949 banquet in San Francisco.

Any member interested may secure one of these pins at a cost of $1.50.

Pins in 18 kt. gold, as well as in platinum, may be had at special prices.

To order, direct request and check to:

Mr. H. A. Sawin,
Yuba Manufacturing Co.,
San Francisco 4, Calif.

New York Ladies to Entertain Husbands

The members of the New York Section of the Woman's Auxiliary to the AIME will entertain their husbands on the evening of Friday, May 13, at the Engineering Woman's Club, 2 Fifth Ave. Cocktails will be served at 6 p.m., followed by dinner and entertainment, all for $3.50 per person.

Among the Student Associates

Louisiana State University

The Student Chapter at LSU has been reactivated, following a wartime cessation of activity. The first meeting, on Feb. 24, was attended by almost 300 students and business men, who heard Francis Anderson, of the Halliburton Oil Well Cementing Co., Duncan, Okla., speak on the use of plastics in unconsolidated sands. The Chapter's officers are James W. Law, President; Jack Simpson, Vice-President; and Jerry Shea, Secretary-Treasurer. H. F. Winham of the Shell Oil Co., Donaldsville, La., has returned as industrial advisor.—*Mahlon F. Manville, Corresponding Secretary.*

Lafayette College

The popular "Story of an Ore Sample" was one of three films shown at the Feb. 22 meeting of the John Markle Society. The Goodrich Rubber Co. contributed "On the Way Out," a film on the use of conveyor belts on coal mines; and the production and use of coal were presented in the last film.

Professor Leist of Lehigh University was there to interest members in the AIME Lehigh Valley Section's essay contest. Sixty of the Society's membership of 75 attended the meeting, later enjoyed refreshments in the ore dressing room.—*J. Siergiej, Secretary.*

University of New Hampshire

Four years of experience with a geophysical crew in the West were presented before the University of New Hampshire's Student Chapter at the January meeting by Harold E. Webster, Jr., geology instructor at the university. Many of the Chapter's June graduates hope to take further courses in oil geology, and so Mr. Webster's discussion of methods, equipment, the work of crew members, and the advantages and disadvantages of this type of work was particularly beneficial to them.

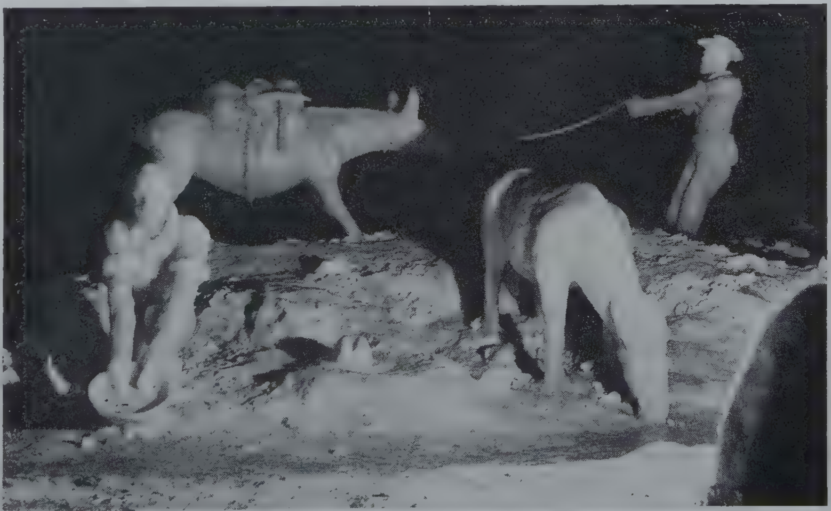
Geologic problems and methods in coal mining, as well as labor problems in the industry, were incorporated in a slide-illustrated talk on coal mining by Ralph Meyers, our Student Chapter's Faculty Advisor, at the February meeting. Professor Meyers' talk emphasized the vast importance of coal in our economy, noting that the petroleum field tended to have a greater place in present day news.—*Marston Chase, Corresponding Secretary.*

University of Southern California

Archibald M. Laurie was elected President of the Student Chapter at USC during the Feb. 16 meeting. Supporting him are Vice-President Richard S. Davis; Secretary Robert J. Fernandes, and Treasurer Robert R. Shaffer. The membership and program committees are headed by Richard Davis and Richard Brazier, respectively. John P. McGriff is the Chapter's parliamentarian.—*Carrol M. Beeson.*

University of Wisconsin

Not nuts and bolts, but a complete delectable dinner, was prepared in the University's metallurgical laboratory furnaces for the Mining Club's February dinner and business meeting. Steeled by their metallurgical meal, the members went on to make plans for St. Patrick's Day observances.



Snow art sculptured by the Mining Club at Michigan College of Mining and Technology took first prize at the winter carnival. The scene commemorates the '49 Gold Rush.

The repast was served at the school's new foundry, and guest speaker Chester V. Nass grasped a fine opportunity to review the past and present in foundries and outline some of the engineering problems extant in that field. Mr. Nass, vice-president of the Pettibone Mulliken Corp., Chicago, and Student Counselor for the Chicago Section, AIME, also reviewed labor-management relations in his industry, and sounded an optimistic note for future job-seekers.—*Arnold D. Arnaut, Public Relations.*

Yale University

Richard J. Christine and I. J. Hook, both of the American Brass Co., addressed the Yale Metallurgical Society on Dec. 2 and Feb. 15, respectively. Mr. Christine told members what industry expects of newly graduated metallurgists; Mr. Hook discussed Anaconda copper production in Montana, and showed a movie on the subject. Discussion periods followed both meetings and proved to be of considerable worth to the students. —*Voluntine T. Potter, Secretary.*

What Went on at Recent Local Section Meetings

SECTION	DATE	PLACE	PRESIDING OFFICER	ATTENDANCE	SPEAKER, AFFILIATION, AND SUBJECT
Arizona, Morenci Subsection	Mar. 7	Longfellow Inn, Morenci	L. O. Ormsby	30	L. L. McDaniel, mill superintendent, Phelps Dodge Corp., on AIME Annual Meeting. L. O. Ormsby, mine superintendent, Phelps Dodge Corp., on trucks for mine haulage.
Black Hills	Mar. 24	Lead, S. Dak.	Fremont Clarke	52	Fremont Clarke, Lithium Corp. of America, on AIME Annual Meeting.
Boston	Mar. 7	Smith House, Cambridge	George P. Swift	48	Frank R. Milliken, National Lead Co., on Ilmenite, Titanium Dioxide and Titanium. Carle R. Hayward, MIT, on R. H. Richards—reminiscences.
Chicago	Feb. 2	Chicago Bar Association	L. R. Berner	125	S. J. Creswell, Carnegie-Illinois Steel Co., on developments in the Bessemer Process.
Chicago	Mar. 2	Chicago	H. A. Herder	54	J. R. Van Pelt, Jr., Battelle Memorial Institute, on tools for modern prospectors.
Colorado	Feb. 23	University Club, Denver	M. I. Signer	34	M. I. Signer, Dean, Colorado School of Mines, on mineral industry education.
Columbia	Mar. 4	Davenport Hotel, Spokane	Howard P. Sherman	24	C. Y. Garber, chairman, Columbia Section, AIME, on AIME Annual Meeting.
Detroit	Mar. 21	Detroit	Frederick P. Bens		Harold Hunsicker, Aluminum Co. of America, on precipitation hardening in aluminum alloys.
Detroit	Feb. 21	Detroit			R. K. Hopkins, M. W. Kellogg Co., on the Kellogg Process of ingot production.
El Paso Metals	Mar. 9	Hotel Cortez, El Paso	E. M. Tittman	26	F. E. Hewitt, asst. prof. of education, Texas Mines and Met., film taken in Russia, 1936.
Gulf Coast	Feb. 15	YMCA, Houston	E. B. Miller, Jr.	122	Olin G. Bell, Humble Oil Co., on deep drilling in the Gulf Coast area.
Lehigh Valley	Dec. 6	Bethlehem, Pa.	M. L. Fuller	104	W. E. Wrather, director, USGS, on recent trip to the Near East.
Mid-Continent	Mar. 14	Bartlesville, Okla.	John P. Hammond	151	Herman H. Kaveler, Philips Petroleum Co., on cases of increasing well densities.
Montana	Mar. 1	Butte	J. Hollis McCrea		H. G. Satterthwaite, Anaconda Copper Mining Co., on the AIME Annual Meeting.
North Pacific	Feb. 17	Rose's Highway Inn, Seattle	W. C. Leonard	39	H. K. Benson, emeritus professor of chemistry, Univ. of Washington, on agricultural production problems.
North Pacific	Mar. 17	Rose's Highway Inn, Seattle	Kenneth Anderson	48	J. Harveson, Standard Oil Co., of California, motion pictures, "Desert Venture"; "City of Destiny"; "Winter Olympiad."
Oregon	Jan. 21	Nortonia Hotel, Portland	Hollis Dole	23	William E. Caldwell, Oregon State College, on northwest strontium mineral deposit.
Oregon	Feb. 25	Mallory Hotel, Portland	Hollis Dole	22	James H. Jensen, Pacific Carbide and Alloys, on calcium carbide for Pacific northwest industries.
Oregon	Mar. 18	Mallory Hotel, Portland	Hollis Dole	31	Pierre R. Hines, consulting engineer, on conveyors and materials handling.
Pittsburgh	Mar. 7	Mellon Institute of Industrial Research	Stephen Krickovic	300	C. H. Snyder, president, Sunnyhill Coal Co., on Colmol continuous mining machine.
Southern California	Feb. 18	Los Angeles	E. O. Slater	70	Richard W. Smith, U. S. Chamber of Commerce, "Government in Your Hair."
Southwest Texas	Mar. 9	Corpus Christi	Houston Parrish	166	George Howard, Stanolind Oil & Gas Co., on proper cementing of casing.
St. Louis	Feb. 18	Hotel York	H. A. Neustaedter	52	Frank G. Breyer, Singmaster & Breyer, on exporting American know-how.
Utah	Feb. 24	Salt Lake City	J. C. Landenberger, Jr.	56	C. J. Christensen, Dean, School of Mineral Industries, University of Utah, on prospectors and Indians. Elroy Nelson, Univ. of Utah, on local taxation in mining industry.
Utah	Mar. 17	Newhouse Hotel, Salt Lake City	J. C. Landenberger, Jr.	44	D. C. Houston, Kennecott Copper Co., "Personnel is People."

Current Institute Papers

The following papers are available to Institute members in the three journals. Each member will receive one classification in the journal of his choice. Members may obtain papers in the two remaining classifications by purchasing the indicated issue of the journal at $0.75 per copy. There are no separate copies of the individual papers. ME means Mining Engineering, JM Journal of Metals, and JPT Journal of Petroleum Technology.

Mining Papers

The Heavy Mineral Deposits of the East Coast of Australia—By N. H. Fisher. (*November Mining Technology.*)

Underground Anemometry — By C. M. Smith. (*January Mining Engineering.*)

Beneficiation of Industrial Minerals by Heavy-media Separation—By G. B. Walker and C. F. Allen. (*January Mining Engineering.*)

Importance and Application of Piezo-Electric Minerals—By Hugh H. Waesche. (*January Mining Engineering.*)

Texas White Firing Bentonite—By F. K. Pence. (*January Mining Engineering.*)

The Application of Screening and Classification for Improved Fine Anthracite Recovery—By W. Julian Parton. (*February Mining Engineering.*)

Operational Statistics of a Marion 5560 Power Shovel—By G. B. Clark and G. L. Reiter. (*February Mining Engineering.*)

The Flotation of Copper Silicate from Silica—By R. W. Ludt and C. C. DeWitt. (*February Mining Engineering.*)

Recent Trends in Asbestos Mining and Milling Practice—By M. J. Messel. (*February Mining Engineering.*)

A Technical Study of Coal Drying—By G. A. Vissac. (*February Mining Engineering.*)

Metals Papers

Transient Plastic Deformation—By R. P. Carreker, J. G. Leschen, and J. D. Lubahn. (*September Metals Technology.*)

Oxide Films Formed on Metals and Binary Alloys. An Electron Diffraction Study—By J. W. Hickman. (*December Metals Technology.*)

Development of the Modern Zinc Retort in the United States—By H. R. Page and A. E. Lee, Jr. (*February Journal of Metals.*)

The Effect of Strain-temperature History on the Flow and Fracture Characteristics of an Annealed Steel—By E. J. Ripling and G. Sachs. (*February Journal of Metals.*)

Relation Between Chromium and Carbon in Chromium Steel Refining—By D. C. Hilty. (*February Journal of Metals.*)

The Densification of Copper Powder Compacts in Hydrogen and in Vacuum—By Charles B. Jordan and Pol Duwez. (*February Journal of Metals.*)

Influence of Composition on the Stress-Corrosion Cracking of Some Copper-base Alloys — By D. H. Thompson and A. W. Tracy. (*February Journal of Metals.*)

Calendar of Coming Meetings

MAY

2 Pittsburgh Section, AIME. President L. E. Young.
2 Boston Section, AIME.
2-4 ASME spring meeting, New London, Conn.
2-5 American Foundrymen's Society, 53rd annual meeting, St. Louis.
5 Reno Branch, Nevada Section, AIME.
6 Columbia Section, AIME.
6-7 American Institute of Chemists, annual meeting, Chicago.
7 Chicago Section, AIME. President's Night.
8-12 AICE, Tulsa, Oklahoma.
9 Mid-Continent Section, AIME.
9-12 American Mining Congress, coal convention and exposition, Public Auditorium, Cleveland.
10 East Texas Section, AIME.
11 El Paso Metals Section, AIME.
11 San Francisco Section, AIME.
11-13 Engineering Institute of Canada annual meeting with International Management Congress hemisphere meeting, Chateau Frontenac, Quebec.
14 Northwest Industrial Minerals Conference under auspices of Columbia Section, AIME, Spokane.
16 Detroit Section, AIME. A. A. Cooley on metallic sheet.
May 16-June 3 U.N. Scientific Conference on Conservation and Utilization of Resources, New York.
17 Gulf Coast Section, AIME.
17 Washington, D. C., Section, AIME.
18 Southwest Texas Section, AIME.
19 Carlsbad Potash Section, AIME.
19 North Pacific Section, AIME.
19 Utah Section, AIME.
19-20 Lake Superior Mines Safety Council, Duluth, Minn.
20 Oregon Section, AIME.
23 Alaska Section, AIME.
24 Montana Section, AIME.
25 Iron and Steel Division, AIME, annual May dinner, Statler Hotel, New York City.
25-26 American Iron and Steel Institute, general meeting, New York City.

JUNE

3-4 Central Appalachian Section, AIME, spring meeting, Blacksburg, Va.
5-10 SAE, summer meeting, French Lick Spring, Ind.
16 Engineering Foundation, 29 W. 39 St., New York City.
June 27-July 1 ASTM, 52nd annual meeting, Chalfonte-Haddon Hall, Atlantic City, N. J.
20-22 American Society of Heating and Ventilating Engineers, semi-annual meeting, Minneapolis, Minn.
20-24 AIEE, summer general meeting, New Ocean House, Swampscott, Mass.
20-24 National Society for Engineering Education, RPI, Troy, N. Y.
25-26 American Iron and Steel Institute, general meeting, N. Y. C.
27-30 ASME, semi-annual meeting, San Francisco.

JULY

9-23 Fourth Empire Mining and Metallurgical Congress, Great Britain.
13-15 American Society of Civil Engineers, summer convention, Mexico City.
15-24 Pan-American Engineering Congress, Rio de Janeiro.

AUGUST

23-26 AIEE Pacific general meeting, Fairmont Hotel, San Francisco.
26-Sept. 10 Engineering & Mining Exhibition, London, England.

SEPTEMBER

25-28 Regional Meeting, AIME, Neil House, Columbus, Ohio.
25-28 American Mining Congress, Western fall meeting, Spokane.
29-30 ASME, fall meeting, Erie, Pa.
29-Oct. 1 Colorado School of Mines, 75th anniversary celebration.
Sept. 30-Oct. 1 Southern Ohio Section of Open Hearth Committee, AIME, fall meeting, Deshler-Wallick Hotel, Columbus.

OCTOBER

5-7 Petroleum Branch, AIME, fall meeting, Plaza Hotel, San Antonio, Texas.
17-19 Institute of Metals Division, AIME, fall meeting, Allerton Hotel, Cleveland.
17-23 AIEE, 1949 Mid-West meeting, Netherlands-Plaza, Cincinnati.
20-21 Petroleum Branch, AIME, Elks Club, Los Angeles.
26-27 Joint Fuels Conference, ASME-AIME, French Lick Springs Hotel, French Lick, Ind.
28 Pittsburgh Section of Open Hearth Committee and Pittsburgh Section, AIME, annual fall meeting, William Penn Hotel, Pittsburgh.

NOVEMBER

1-5 Pacific Chemical Exposition, California Section, American Chemical Society, San Francisco Civic Auditorium.
2-4 American Society of Civil Engineers, fall meeting, Washington, D. C.
7-10 AIChE, annual meeting, Pittsburgh, Pa.
9-11 Industrial Minerals Division, AIME, Tampa, Fla.
12-14 Geological Society of America, annual meeting, Hotel Cortez, El Paso.

DECEMBER

8-10 Seventh Annual Conference, Electric Furnace Steel Committee, Iron and Steel Division, AIME, Hotel William Penn, Pittsburgh.

FEBRUARY 1950

12-16 Annual Meeting, AIME, Statler (Pennsylvania) Hotel, New York City.

APRIL 1950

10-12 Open Hearth Conference, and Blast Furnace, Coke Oven and Raw Materials Conference, Nethland-Plaza Hotel, Cincinnati.

DECEMBER 1950

7-9 Eighth Annual Conference, Electric Furnace Steel Committee, Iron and Steel Division, AIME, Hotel William Penn, Pittsburgh.

THE DRIFT OF THINGS

. . . as followed by EDWARD H. ROBIE

For Peace or War?

Unquestionably, the North Atlantic Pact is an important milestone in international agreements destined to prevent war, but whether or not it will be more effective than previous treaties remains to be seen. The approach is quite different from the Briand-Kellogg treaty—the Pact of Paris—signed by fifteen nations in 1928, and subscribed to by others including Russia later. That was an out-and-out renunciation of war, which certain of the signers, notably Germany and Japan, did not take seriously, as subsequent events proved. A different type of agreement designed to prevent war was that embodied in the League of Nations, and the Permanent Court of International Justice, and latterly in the United Nations. A weakness in the last-named, of course, is the privilege of the veto power, which both Russia and the United States insisted on reserving and which Russia has used to excess. No joint action, such as the United Nations agreement implies, can work if one nation reserves the right to prevent action in cases where its own interests are involved. Another weakness is lack of police power to enforce its decisions. Thereby it can become little more than an international debating society. The world government movement, as is being conducted with some success by United World Federalists, would seem to be a much more realistic means of preventing war. Soviet Russia will have none of it, but might subscribe to it if the rest of the world did.

Heretofore military alliances have been notably unsuccessful in preventing war, as have all other means, but at the present juncture in world affairs, with Soviet imperialism rampant, some forceful repressive measures seem necessary. Russia has long shown an unco-operative spirit with the rest of the world. Appeasement gives no promise, having failed so signally with another totalitarian government imbued with the idea of re-forming the world.

The North Atlantic Pact might better serve its purpose of scaring off aggressors, and preventing war, if more nations joined it. We already have the treaty of Rio de Janeiro, signed in 1947, whereby the American nations pledge themselves to unite against any one that starts a war. This would seem to fit in with the new pact. If, now, all the rest of the world would enter into such a treaty and aggression were sharply defined, then the force that the United Nations lacks to back up its decisions would be provided and a large part of the aims of the world government advocates would be realized.

Russia's opposition to the present pact means one of two things—either that she herself has designs on still more neighboring countries, or that she doubts that the pact is designed merely to stop aggression. In view of her inexcusable attack on Finland before World War II, and her seizure of control of Czechoslovakia after the war, to cite but two examples, one may be pardoned for considering the first reason the paramount one.

Therefore, a military alliance against aggression, such as the North Atlantic pact, seems well justified at the present time and likely to be a deterrent to war, even though it may not prevent one. But the entire world should join such a pact—and then live up to it. In other words, the United Nations with a gun in its pocket.

We have no sympathy whatever with the minority that think Russia will do as she pleases as soon as she gets the atomic bomb and that therefore we should dispose of her as soon as possible. Not only would that be morally reprehensible, but the practical result of another war so soon after the last might easily be the overthrow of our system of government. It has already been undermined by two disastrous world wars, and if we cannot prevent another, then it is high time something different should be tried. Not that Communism gives hope of being anything better.

The best rule for the present, as we see it, is that to which Theodore Roosevelt gave voice: "Speak softly but carry a Big Stick." The Big Stick in this case is the North Atlantic Pact, with the requisite power to carry it out.

Encouragement for Our Teachers

Winston Churchill, in his fine address given at the Mid-Century Convocation of MIT, paid a well-deserved tribute to our American engineering schools: "We have suffered in Great Britain by the lack of colleges of university rank in which engineering and the allied subjects are taught. Industrial production depends on technology and it is because the Americans, like the prewar Germans, have realized this and created institutions for the advanced training of large numbers of high-grade engineers to translate the advances of pure science into industrial technique—it is for that reason that their output per head and consequent standard of life are so high."

He might have added that for the same reason this country was the first to produce enough fissionable material to make possible the practical use of nuclear energy.

Our engineering schools have long been the target of critics who have complained of inadequate training. This view of a foreign observer should give the schools much encouragement.

Dr. Boyd Confirmed As Director

Mining men were glad to hear that Jim Boyd had finally been confirmed as director of the Bureau of Mines and no longer had to serve without a salary. In fact, we understand that his back salary is to be paid. Unfortunately, the directorship of the Bureau has for some years depended more than it should on politics. But fortunately Dr. Boyd is a comparatively young man, and able professionally, so perhaps, with a good record behind him, he can survive any change in the political complexion of the Administration, should one ever occur.

The delay in confirming Dr. Boyd as director is said to have been due to the opposition of John L. Lewis, who has no use for Secretary Krug, Dr. Boyd's boss. John L. was also able to bring political influence on Senator Millikin, of Colorado, who long was able to block the appointment but was finally overruled in the new Administration. When Lewis called the two-week strike of the coal miners, purportedly to call attention to the need for safer working conditions, the matter was brought to a head, and the Senate confirmed Dr. Boyd partly to show its independence of Lewis. Even so, eleven Senators voted against the appointment, with fifty favoring it. Perhaps the names of the eleven should be remembered. They were: Millikin, of Colorado, Brewster of Maine, Bricker of Ohio, Capeheart and Jenner of Indiana, Langer of North Dakota, Schoeppel of Kansas, and Wherry of Nebraska, all Republicans; with Johnson of Colorado, and Kilgore and Neely of West Virginia, Democrats.

The principal argument offered against Dr. Boyd was that he had not had operating experience in mining, and had had no contact with either the coal mining industry or with health and safety matters. Operating experience would not appear necessary to fill the job acceptably; and men with both coal and metal-mining experience combined with health and safety training are exceedingly rare.

The Accident Record

The real reason for the two-week shutdown of the coal mines was probably not so much as a memorial to coal miners who had been victims of 55,000 coal-mine accidents last year, as alleged, as it was to reduce the stocks of coal a bit, so as to afford better bargaining conditions when the time for setting a new contract comes up, on May 1. The shutdown had little practical effect, for coal demand has recently been much less than capacity, so that a large number of the miners have been working only two to four days a week. Were wages lower, which would permit lower coal prices, then more coal in relation to oil would be used, and it might be that the income of coal miners would be greater than at present. But high wages make the best selling point, even though the opportunity to get them is limited.

If the reason for the shutdown was really the one given, then the timing was not particularly propitious, for the news came out during the strike that in 1948 there were fewer accidents per ton of coal mined than in any previous year in history. As to fatalities, the record was not quite so outstanding, but still good: it was the fourth lowest year as to deaths in bituminous mines and the second lowest in anthracite mines. For both branches, 1010 men were killed last year and 54,100 injured. That is still not a record to be proud of, nor is the record of work injuries for all industries. Last year disabling work injuries amounted to 1,960,000, deaths 16,500, permanent disablements 1800; with 41,000,000 man-days of work lost, representing more than four billion dollars in wages. Automobile fatalities were around 32,000.

It's dangerous to be alive. Carelessness takes a bigger toll of life and health than war but is not nearly so costly in money.

AS&R's Golden Anniversary

On April 4 the American Smelting and Refining Co. celebrated its fiftieth anniversary. Receptions and entertainment were features of the day at many of its plants, and the employes of the headquarters office in New York were entertained at the Bankers' Club. Though the words "mining" and "milling" are not included in its name, its activities include these fields as well, and extend all over the world.

We have been visiting AS&R plants, and have been on friendly terms with AS&R men, for 35 years. We have found the plants well run by a fine lot of men who have been uniformly most generous and hospitable when we dropped around. Until we decided it was more important to buy a home than to be an AS&R stockholder we used to receive regular dividends, of modest amount but an excellent return on our small investment. AS&R men have written innumerable papers for the AIME, and have served on numerous committees, and in various capacities as officers. One is currently a Director.

The Company produces practically all nonferrous metals, to some extent at least, but is important in copper, lead, zinc, gold, and silver. Its interests are in both primary and secondary metals, and extend to fabricating. It operates several coal mines but so far as we know has kept out of the oil business. Much of its raw material is purchased from small shippers but much smelting and refining is done for large companies too, like Kennecott, for instance.

So we add our salute to the others that AS&R has received. For half a century it has conducted important units in a basic American industry with conspicuous success to itself and to its stockholders, and in a highly creditable manner as a national institution. May the next half century add further to its stature.

News of AIME Members

Wayne J. Ahl is a junior engineer with the Inland Steel Co. His address is 336 W. Division St., Ishpeming, Mich.

L. W. Allen, on six months' leave from his post as underground manager of Nchanga Consolidated Copper Mines Ltd., Chingola, N. Rhodesia, spent the first three months of that leave traveling some 6000 odd miles, doing a little fishing here and there, taking in the bright lights of Johannesburg and one or two other cities in S. Africa, and is spending the last three months on his farm, quite close to the coast in a beautiful part of South Africa. The fishing is good, and with the farm to potter about on, he expects to enjoy himself. Nchanga Consolidated is still busy with an expansion program and has a big construction and development program on hand. The tonnage mined has been steadily increasing and just before he left had reached a record of some 114,000 tons a month. The first increase in capacity is designed to handle 150,000 tons a month, but it looks as if the plant will have a rather larger capacity than that for which it was designed.

Edgar F. Andersen, formerly with the Golden Queen Mining Co., is now working for the Calaveras Cement Co., San Andreas, Calif.

Alfred L. Anderson, professor of economic geology at Cornell University, Ithaca, N. Y., is spending his Sabbatical leave of absence in the spring term visiting mining camps in the Rocky Mountain region.

Lloyd E. Antonides receives his mail at 13 Church St., Franklin, N. J. He is surveyor for the New Jersey Zinc Co.

Florentino B. Baca has been plant manager of the Hardsocg Mfg. Co., manufacturers of power driven rotary drilling equipment and miners tools and supplies, a subsidiary of the Cardox Corp. of Chicago, since November, 1947. He lives at 718 W. 4th, Ottuma, Iowa.

Wayne O. Axtell

Wayne O. Axtell, Colonel in the Corps of Engineers, has retired from the Army after a final tour of duty in the Pacific. He is now located as an engineer in the office, Chief of Engineers, Washington, D. C.

W. R. Allen

W. R. Allen is chairman of the board of directors of the Bostana Mines Co. and consultant for Elkhorn Beaverhead Mines Co. He has on the press, ready for distribution, a story of western development and particularly early day mining history in Montana with a backdrop of national events and personalities.

Howland Bancroft, mining consultant, of New York, who has spent the past eighteen months on professional work in South and South West Africa with headquarters in Tsumeb, returned to the States early in March. He will be engaged in an investigation in this country for the next few months, following which he will return to South Africa about mid-year, with headquarters in Johannesburg.

Robert W. Barnes can be reached at 506 Nancy St., Charleston, W. Va. He is employed by the Warner Collieries of Mammoth, W. Va.

Henry E. Bartony, Box 231, Nellis, W. Va., is employed in the engineering department of the Armco Steel Corp.

Andre C. Beguin, who had been a student at the University of British Columbia, is mill engineer for Torbrit Silver Mines, Ltd., Alice Arm, B. C.

Charles H. Behre, Jr., vice-president of Behre Dolbear and Co., New York City, left the States recently for Burma. The Company has been appointed general consultants and technical advisers on minerals for the Burmese Government. Dr. Behre will make his headquarters at Rangoon and prepare recommendations outlining the scope of industrial and mining rehabilitation.

John R. Bogert is employed as a field geologist by the Bethlehem Steel Co., and is addressed in care of W. A. Rankin, construction superintendent, Iron Mines Co. of Venezuela, San Felix, Est. Bolivar, Venezuela.

John P. Bowman is employed by the Foster Wheeler Corp., Carteret, N. J., as a service engineer.

W. A. Broughton is head of the geology department at the Wisconsin Institute of Technology, Platteville, Wis. His address is 225 Bradford, Platteville.

Donald M. Brown is working with the U. S. Geological Survey, estimating coal reserves. His address is 19th and F Sts., FWA Bldg., Washington 25.

Frank A. Burns, who was a student at Lehigh University, is working for the Bethlehem Collieries Corp., Johnstown, Pa., as a junior engineer. His mailing address is Box 62, Ellsworth, Pa.

John L. Beach is chief engineer-geologist in charge of exploration for Rhodesia Copper Ventures Ltd., P. O. Sinoia, S. Rhodesia.

E. L. Beck and his associates have taken over the operation of the Perch hydraulic mine at Orleans, Calif.

Edward J. Bottomley is superintendent of the Sunlight Coal Co., Boonville, Ind. He was with the Tecumseh Coal Corp.

Harold L. Brandt has taken the post of assistant engineer with the Pend Oreille Mines and Metals Co., Metaline Falls, Wash. He had worked in Chihuahua, Mexico, for the American Smelting and Refining Co.

Willis F. Bronkie is vice-president and secretary of the Rising & Nelson Slate Co., West Pawlet, Vt. He has been sales engineer for the Victaulic Co. of America.

Alan D. Brown, graduated from the University of Minnesota School of Mines last June and is now employed by the Anaconda Copper Mining Co. in Butte, Mont.

David W. Boise, formerly of Hurley, N. Mex., where he was working with the Kennecott Copper Co., now lives at 917 West St., Silver City.

Franklin W. Butterfield has transferred from the University of Nevada to the Placer College, School of Mines, Auburn, Calif. He is employed part time by the Empire Star Mines Co. and receives mail at P. O. Box 817, Grass Valley, Calif.

Sam S. Coldren has resigned from the Acoje Mining Co., Santa Cruz, Zambales, to become manager of the Misamis Chromite Co., Na-awan, Initao, Misamis Oriental, P. I. The Company is operated by International Engineering Co. Warner-Barnes.

Alden P. Colvocoresses writes from Box 537, Superior, Ariz., that he has taken over and is maintaining the files of his late father, **George M. Colvocoresses.** These files contain reports and data on approximately 1000 different mining properties in Arizona and adjacent states.

Annan Cook has given up his post as chief geologist to Frobisher Ltd., in Africa, and has accepted a job in the exploration department of the Kennecott Copper Corp.

Boyd Appointment Confirmed

James Boyd's appointment as director of the Bureau of Mines was confirmed by the Senate on March 22 by a vote of 50 to 11. Here he is sworn in by Chief Clerk Dotson as Secretary of the Interior J. A. Krug looks on. The former dean of the Colorado School of Mines took this post in August, 1947, and received pay for about four months until a short session of Congress adjourned in December of that year without confirming him. John L. Lewis, president of the United Mine Workers, who had been blocking confirmation for more than two years, was mentioned only once in the hour-long debate in the Senate and the coal mining stoppage called by Mr. Lewis, partially in protest against Dr. Boyd's appointment, was not mentioned at all.

Ronald Crawford lives at Apt. D., 2509 Park Lane Court S., Mountain Brook, Ala., while he works for the Equipment Service Co., Birmingham, as a sales engineer.

W. H. H. Cranmer, president and general manager of the New Park mining Co. of Keetley, Utah, succeeds **James Ivers,** vice-president and general manager of the Silver King Coalition Mines Co., as president of the Utah Mining Association.

L. A. Dahners has traveled from Anchorage, Alaska, to Skouriotissa, Nicosia, Cyprus, and can be reached there in care of the Cyprus Mines Corp.

James R. Diehl is now employed in the technical department of the engineering and construction division of the Koppers Co. at Pittsburgh. His home is at 409 N. Neville St., Pittsburgh 13.

Ira F. Davis, who is a consulting mining engineer, is now living on Meadville Dr., in Onancock, Virginia.

Byron M. Emery recently graduated from the Pennsylvania State College with a B.S. degree in mining engineering and is employed by the Atlas Powder Co., Wilmington 99, Del., as a technical representative.

John W. England is assistant geologist for the Colorado Fuel and Iron Corp., Pueblo, Colo. His address in Pueblo is Apt. 604-K Air Base.

Harry J. Evans, mining engineer of the Pierce Management staff in Scranton, Pa., left for Korea on March 9 to study the economic potentialities of four coal fields in southern Korea.

A. L. Fairley, Jr., has been made assistant to the president of the Snyder Mining Co. which operates three iron mines on the Mesabi range of Minnesota.

John H. Ffolliott, formerly with the Tennessee Copper Co., is now with the Miami Copper Co., Rm. 1824, 61 Broadway, New York City 6.

John W. Foreman is working for the Monroe Coal Mining Co., Revloc, Pa., as a junior mining engineer.

Edward S. Frohling, after graduating from MIT, worked for a brief period as research engineer for the Division of Industrial Co-operation there. He is now employed by the Climax Molybdenum Co. as a metallurgical testing engineer.

Raymond B. Gallant is now with E. I. du Pont de Nemours & Co. in the explosives sales division. He represents the Company out of the Denver, Colo., office and covers northeast Wyoming and central Montana.

Conrad I. Gardner has gone to Potosi, Bolivia, as engineer for Cia. Minera Unificada del Cerro de Potosi, Casillas 30 y 52.

P. W. Gifford, Jr., who had been general mine foreman for the Lone Star Steel Co., is now superintendent of the slag crusher of Gifford-Hill & Co., Daingerfield, Texas.

B. E. Grant, since 1944 general superintendent of the U. S. and Lark mine of the U. S. Smelting Refining and Mining Co., has been promoted to assistant to the manager of Western mines succeeding **A. G. Kirkland** who became assistant manager of Western mines on Jan. 1. **M. M. DuBois** has been made general superintendent of the U. S. and Lark mine and **Harold H. Wells,** former assistant general mine foreman of the U. S. section, has become assistant superintendent of the Lark section.

Allen D. Gray, Jr., is taking the training program of the Rochester & Pittsburgh Coal Co., Indiana, Pa. His address is 1058 Church St., Indiana.

William A. Griffith and **William A. Griffith** were mixed up in a personal item in the March issue. The former is an instructor at MIT, with an address at 81 Marion St., Brookline, Mass., while the latter is a student at the University of Idaho, Moscow, according to our records. We apologize to the Mr. Griffiths and hope our error has caused them no inconvenience.

F. W. Goddard has been transferred by the AS&R Co. from their Avalos mill unit at Chihuahua to their Parral unit at Parral, Chih., Mexico, in the same post of assistant mill superintendent. The Parral mill produces concentrates of lead, zinc, and copper, milling 1500 tons a day.

Mostyn G. Grant is working for the Coronado Copper and Zinc Co., Los Angeles. His home is at 2031 Pelham Ave., West Los Angeles.

Harry G. Grigsby, former Missouri School of Mines student, is working for the Idarado Mining Co. at Ouray, Colo.

Ivan D. Hagar retired from the Titanium Division of the National Lead Co. last May and moved to R. F. D. 2, Middlebury, Vt. He is retained by National Lead as a consultant.

D. F. Hewitt has left the University of Wisconsin and joined the Ontario Department of Mines staff, Parliament Bldg., Toronto 2, Ont.

Arthur Horen, formerly assistant geologist for the Lone Star Steel Co., is now pursuing graduate study in geology at Harvard. His new address is 27 Maple Ave., Cambridge 38, Mass.

Roy Hatch, labor relations director at the Utah Copper Division of the Kennecott Copper Corp., has been advanced to assistant to the general manager with jurisdiction over all employe and public relations. **L. F. Pett,** assistant mine superintendent, has been made mine superintendent, and **J. C. Landenberger, Jr.,** formerly general mill foreman, replaces Mr. Hatch as director of labor relations.

George Deike and Eugene Jones

Eugene H. Jones, majoring in coal mining engineering at the University of Pittsburgh, received the Old Timers Club award from **George H. Deike,** president of Mines Safety Appliances Co. The Old Timers Club, a social organization of leading coal mining executives, recently decided to give prize awards to outstanding seniors majoring in coal mining engineering at several schools of mines. On Feb. 10 Mr. Deike awarded a gold watch to Mr. Jones, whom a committee had decided upon as outstanding in scholarship, character, and promise. The watch was engraved "Old Timers Club Award to Outstanding Senior Student, Eugene H. Jones, University of Pittsburgh, Class of 1949, Coal Mining Engineering."

F. S. McNicholas and Mrs. McNicholas

F. S. McNicholas returned to Denver the first of April following a month of consulting work on block caving methods at the Mufulira Copper Mines property in Northern Rhodesia. He was accompanied by Mrs. McNicholas.

A. L. Hayes, former superintendent of the Empire Zinc Co. at Hanover, N. Mex., has been transferred to the New York office of the New Jersey Zinc Co., where he will assist the general manager of mines, **S. S. Huyett** has been appointed superintendent of the Empire Zinc Co. at Hanover.

Ira K. Hearn, Jr., who was formerly works industrial engineer of ore mines and quarries for the Tennessee Coal, Iron and Railroad Co., Bessemer, Ala., has recently accepted the post of assistant to the president of the Quebec Iron and Titanium Corp., a subsidiary of Kennecott Copper. His office is at 120 Broadway and he is making his home at the Park Garden Apartments, 134-42 Franklin St., Morristown, N. J.

Arthur S. Hecht, consulting engineer of San Francisco, is now an associate of Behre Dolbear and Co., consulting mining engineers and geologists, 11 Broadway, New York City 4. Mr. Hecht recently returned from Korea and completed an assignment for the Civil Affairs Section of the Army and the ECA for Korea as a consultant, in the United States, on a research project for tungsten ore beneficiation.

Charles T. Holland has joined the staff of the mining engineering department of the Virginia Polytechnic Institute. He had been professor of mining engineering at West Virginia University. Mr. Holland succeeds **L. I. Cothern** who is head of the mining engineering department at Ohio State University.

Fred H. Howell is employed as a mining engineer and geologist with the Federal Mining and Smelting Co. at the Page mine, Page, Idaho.

Charles O. Hower is addressed at Emmett, Idaho, and works as metallurgist for the Calera Mining Co. at Forney.

Charles D. Hoyt graduated from Penn State on Jan. 31 and is now employed by the Westmoreland Mining Co. His new mailing address is 143 S. Spring St., Blairsville, Pa.

Lemuel B. Hunter has been made manager of the raw materials department of the Inland Steel Co., Chicago.

Nick Hurley has resigned his post with Cie. Aramayo de Mines en Bolivie at La Paz and can now be reached at 1510 Mathews, Fort Collins, Colo.

Robert H. Inman has taken the job of assistant engineer with the United Electric Coal Co., Chicago. Mail reaches him at 403 N. First St., Canton, Ill.

John A. Johnson is acting supervisor at the Duluth office of the Bureau of Mines.

W. Bruce Johnson is employed as a coastal analyst for the Navy Department. His address is Rm. 3A-538, Pentagon Bldg., Washington 25, D. C.

Russell C. Johnson has changed his address to in care of the Northern Peru Mining and Smelting Co., Casilla 219, Lima, Peru. He had been with Cia. Minera Asarco.

Theodore L. Johnston has resigned as metallurgist for the Bureau of Mines to do consulting work with headquarters at Hamilton, Mont.

Frederick C. Kruger, assistant chief geologist of the Cerro de Pasco Copper Corp., is on leave of absence and will be at Northwestern University, Evanston, Ill., as lecturer in geology during the winter and spring quarters.

Thomas A. King, Box 651, Clarksville, Pa., is a student engineer with the Westmoreland Mining Co., Blairsville, Pa.

Charles P. Knaebel and a partner have taken a lease from the U. S. Smelting Refining and Mining Co. on its Shingle Canyon mine, near Fierro, N. Mex. The operations consists of small scale production of zinc-lead ore.

Edward H. Kraus has been, since February, 1945, on the retired list of the University of Michigan. However, this does not mean that he is not pursuing actively mineralogical and gemological affairs. In fact, with his colleague, C. B. Slawson, they were able to bring out the fifth edition of Gems and Gem Materials in December, 1947. At present, with W. F. Hunt and L. S. Ramsdell, he is at work revising the Mineralogy and hopes to bring out the fourth edition within a year or two.

H. M. Lavender has been made a member of the board of directors of the Phelps Dodge Corp. He is vice-president and general manager of Phelps Dodge in charge of all western operations of the Corporation.

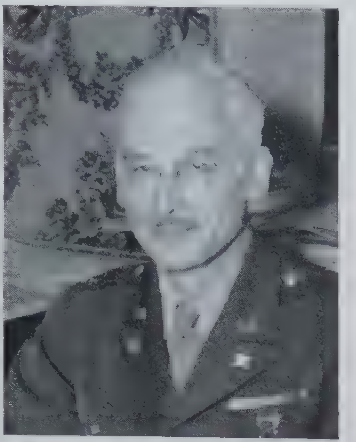
John W. Lowry is process metallurgist in the division of metallurgical research of the Permanente Metals Corp., Spokane, Wash. Mail reaches him at E. 10918 Main, Opportunity, Wash.

John R. LeGrand is working for the Atlantic Coast Line Railroad, Wilmington, N. C., as an industrial geologist.

Robert T. Macaul is working in the rock drill department of the Ingersoll-Rand Co., 11 Broadway, New York City 4.

Edward Marks, while working on his Master's degree, is a student assistant in the department of geology at the University of Texas, Austin.

Victor R. Martin lives at 642 S. Gilpin St., Denver, and during working hours is field engineer for the National Fuse and Powder Co. He moved to Denver from Butte, where he was underground engineer for the Anaconda Copper Mining Co.



Charles Gill Morgan

Charles Gill Morgan returned to Dallas, Texas, in January and has offices in the Tower Petroleum Bldg., specializing in petroleum geology and research in seismology and glaciology. After graduation from SMU, he served with the Geophysical Research Corp. and the Seismograph Service Corp., spent two years in graduate study at Harvard, and then was off to the Antarctic with Admiral Byrd as chief geologist on his second expedition. In 1936 he became associated with Herbert Hoover, Jr., as vice-president of the United Geophysical Co. in California. Called to active duty in Washington as a Captain in the Air Corps in 1941, Colonel Morgan was successively in charge of Latin American intelligence, European targets, and finally the Operational Intelligence Division for both theaters of war. As the war neared its close in Europe, he was relieved from his assignment by the War Department in Washington to serve on the USAF Evaluation Board as director of tactical air operations in Europe. He returned to inactive duty in March, 1946, and has been engaged in independent research and as consultant in Boston and New York since that time.

John F. Marley has been assistant mine superintendent with the Gouverneur Talc Co. at Balmat, N. Y., since last October. His address is 179 Rock Island St., Gouverneur, N. Y.

Richard R. Matthew, county surveyor and mining engineer for the U. S. Grant Mining Co., who lives in Virginia City, Mont., is the president of the Madison County Mining Association.

Arne A. Mattila has been appointed mining engineer of the mining branch, conservation division, U. S. Geological Survey, and will assist with the supervision of mining operations on leased government land in the Billings, Mont., district. Mr. Mattila is a graduate of the Montana School of Mines, class of 1936. His mining experience includes about seven years with the Anaconda Copper Mining Co. and three years as mining engineer with the Republic Coal Co. of Roundup, Mont.

Walter L. Maxson, vice-president of the Oliver Iron Mining Co., Duluth, Minn., has been appointed to the Atomic Energy Commission's Advisory Committee on Raw Materials. Mr. Maxson has been an officer of the Oliver company since 1944. Prior to that he was sales manager and chief engineer of the mining division of the Allis-Chalmers Mfg. Co. for seventeen years. He has also served as associate professor of metallurgy at the Colorado School of Mines, and has been connected with several mining companies in Colorado, Arizona, and Australia.

John W. Mrock transferred on Jan. 28 from the Bureau of Reclamation, Coachella, Calif., to the Navy Department, and is employed at present as a laboratory technician at the Naval Air Missile Testing Center at Point Mugu, Calif. His new mailing address is 245 W. 1st St., Oxnard, Calif.

Harvey S. Mudd, president and managing director of Cyprus Mines Corp., officer of several West Coast business organizations, and past president of the Institute, was given the Egleston medal, awarded annually by the Columbia University Engineering School Alumni Association for distinguished engineering achievement, on April 28. Mr. Mudd's principal professional activity for many years has been the development and management of important copper mines in the Island of Cyprus. From 1914 to 1926 he devoted his energies to these and other mining properties. During the many years of his connection with Cyprus Mines, Mr. Mudd has solved numerous complex mining and metallurgical problems involving mine ventilation and safety, fire hazards, and production.

Henry D. Mumper is engineer in the steam and combustion department of the Bethlehem Steel Corp., Johnstown, Pa. His mail goes to 711 Graham Ave., Windber, Pa.

John C. Murray, who has been at the Pittsburgh works of the Jones & Laughlin Steel Corp., has joined the staff of the Republic Steel Corp. at Youngstown, Ohio.

Dwight L. Myers recently accepted

a post under the city engineer in Spokane, Wash., as construction engineer. His work will be wholly devoted to a $3,500,000 construction job of which about ten per cent has already been awarded to contract.

W. Clark Mason has resigned as metallurgist for the Inspiration Consolidated Copper Co. and has gone to San Francisco as metallurgist for the Merrill Co. His new address is Hill Castle Apt., 1431 Jackson St., Oakland 12, Calif.

R. K. Matheson is now reached at the Coronado Copper and Zinc Co., Bella Vista, Calif.

Donald H. McLaughlin has been appointed chairman of the National Committee of sponsors of the Harvard Foundation for Advanced Study and Research.

Herbert H. Mellus is pit engineer of the new open pit project of the Inspiration Consolidated Copper Co., Inspiration, Ariz.

James J. Merle has changed his address to the Fairview Collieries Corp., Box 86, Danville, Ill. He was chief chemist for the Ayrshire Patoka Collieries Corp.

Henning C. L. Meyer, who used to be with the New Jersey Zinc Co., has moved to 1507 S. 78th St., West Allis, Wis. He is working for the Allis-Chalmers Mfg. Co.

Gilbert Monroe graduated in mining engineering from the University of Alaska last May and now works as a junior engineer with an aircraft company. His address is Apt. 12, Darrell Hotel, 703 8th Ave., Seattle, Wash.

Perry F. Narten, after graduating from Washington University, St. Louis, Mo., went to work for the U. S. Geological Survey, Washington 25, D. C.

M. J. O'Shaughnessy has become project superintendent of Allard Lake (Quebec) Mines Ltd., Havre St. Pierre, P. Q. He was with the Alger Gold Mines.

W. P. Putnam has retired from active business and is spending his winters at 1001 Hamlin Ave., Orlando, Fla., where he is enjoying a change from northern winters.

Edwin H. Oshier is field superintendent for the Round Mountain Gold Dredging Corp., Round Mountain, Nev. He had been with the Golden Queen Mining Co., Mojave, Calif.

W. C. Page became assistant general manager of Western operations of the U. S. Smelting Refining and Mining Co. on March 1. In addition to his general duties, Mr. Page will continue to have direct supervision over the Midvale plant operations and the refining operations of the Company.

K. Peters

K. Peters, formerly manager in charge of the mining and metallurgical operations of the Mahad Dahab property of the Saudi Arabian Mining Syndicate, has been appointed consultant on mining affairs to the Kingdom of Saudi Arabia. Mr. Peters will make his headquarters in Jedda, Saudi Arabia.

Louis A. Panek is associated with the Bureau of Mines at College Park, Md., doing research in connection with roof control and the use of suspension supports. He is visiting properties where suspension supports are being introduced in order to keep informed of the problems being encountered in the field.

Everett M. Paris is employed as mine superintendent with the U. S. Vanadium Corp. at Uravan, Colo.

Roy I. Petersen is assistant superintendent of the shaft sinking division of the Bethlehem Collieries Corp., and is addressed at R. D. 1, Bentleyville, Pa.

Alan P. Ploesser has gone to Baguio, Philippine Islands, to work for the Benguet Consolidated Mining Co.

O. Edward Pothier is a mining engineer in the exploration department of the Simplot Fertilizer Co., Pocatello, Idaho. He receives mail at Rt. 7, Boise, Idaho.

Emmett S. Pugh, president of the Raleigh Smokeless Coal Co. until 1946 when it was sold to the Blackstone Coal Co., has announced the organization on Feb. 1 of a new operating sales company called the Raven Pocahontas Co. The new company will be owned entirely by the producing companies represented, the officers, and the employes. The four operating companies concerned at present are Raven Coals, Inc., the Raven Peerless Co., Raven Red Ash Coal Co., and an affiliate, and the Red Ash Mining Co., which are producing about 2500 tons per day of mid-volatile coal.

Richard Quirk has been appointed assistant superintendent of the Morococha department of the Cerro de Pasco Copper Corp., Morococha, Peru. He was formerly general mine foreman there.

Alfred L. Ransome has been transferred from his former post as supervising engineer of the San Francisco office of the metal economics branch of the Bureau of Mines to one of similar capacity with the newly established metal economics branch office in Juneau, Alaska. On Dec. 5, a little "sourdough," David Miles, 7 lb. 4 oz., was added to the Ransome family roster by way of St. Anns Hospital, Juneau.

Avery H. Reed, Jr., is employed as a mining engineer with the Bureau of Mines, Box L, University, Ala. He had been general superintendent of the Gilbert & Hageman Co., Lexington, Ky.

Gilbert R. Richardson has gone "down under" to Sydney, N. S. W., where he is reached at the Eveready (Australia) Pty. Ltd., P. O. Box 37, Mascot.

A. E. Roberts, since he was graduated from the College of Mines at the University of Arizona last May, has been employed on the engineering staff of the Magma Copper Co. at Superior, Ariz.

Ernest H. Rose, now doing research and development work for the Tennessee Coal, Iron and Railroad Co., has been appointed to the Committee on Raw Materials of the Atomic Energy Commission. Mr. Rose, whose work with TCI is directed toward the exploitation of Alabama's silicious iron ores, was mill superintendent for the International Nickel Co. of Canada at Copper Ciff, Ont., for many years.

Clyde R. Russell, who was mine superintendent of the Tonopah Belmont Consolidated Mining Co., Tonopah, Nev., has gone to Oruro, Bolivia, where he is addressed at Casilla No. 154.

Margaret A. Ryan went to work for the Anaconda Copper Mining Co. in January as a technical chemist at the Washoe smelter, Anaconda, Mont.

V. P. Sokoloff is a soil scientist and geochemist in the geochemical prospecting unit of the U. S. Geological Survey, Denver Federal Center, Denver 14, Colo.

Alex E. Stutter, secretary of the Committee on Membership, AIME, and his wife Frances, are the proud parents of George Alexander who was born March 18 at the Bethany Deaconess Hospital, Brooklyn, N. Y.

Robert A. Stauffer has been elected vice-president and director of research of the National Research Corp., Cambridge, Mass. Mr. Stauffer, a Harvard graduate, has been associated with the company since 1942.

J. S. Sumner is engaged at present in geophysical prospecting on the Marquette range. He is addressed in care of the geological department of the Cleveland-Cliffs Iron Co., Ishpeming, Mich.

Charles Will Wright has been acting as consulting engineer for ECA on a part-time basis, but his permanent connection continues as consultant to Westinghouse Electric International, Rm. 1810, 40 Wall St., New York City. He spends a part of each week in Washington, where his home address is 2540 Massachusetts Ave., Washington 8, D. C.

William E. Wright has changed his professional address from Pennsylvania State College to commodity specialist, U. S. Tariff Commission, Washington 25, D. C.

George G. Yeager, with the Aluminum Co. of America, has been transferred to the company's purchasing department in Pittsburgh. His address is 3229 Circle Drive, Pittsburgh 27, Pa.

• In the Metals Divisions

Elbert E. Alpers works for the General Motors Corp. central foundry division as a metallurgist. Mail reaches him at Apt. 2, 1201½ E. Main St., Danville, Ill.

J. R. Brady has been made supervisor of the research and testing department of the Wisconsin Steel Works, Chicago.

M. H. Caron, of Delft, and **Herman Ebbinge,** of Den Haag, Netherlands, have recently been in the United States and Canada on professional work. Professor Caron reports that the university at Delft is in serious need of a copy of AIME Transactions volume 121, on lead and zinc, now out of print. If a member has a copy he can spare, he would like to be advised.

E. Corbin Chapman has been appointed chief metallurgist of Combustion Engineering-Superheater, Inc., and will continue to make his headquarters in Chattanooga, Tenn. Mr. Chapman has been employed by Combustion since his graduation from Cornell in 1928 and is the author of numerous articles and papers on welding and metallurgical subjects. During the second World War he served as supervisor of the National Defense Research Council project on development of electrodes for welding armor plate.

Herbert J. Cooper is working for the Cooper Alloy Foundry Co. as an engineer on research and development. He lives at 336 Harding Dr., South Orange, N. J.

Otto Barth

Otto Barth on Feb. 25 was named professor of nonferrous metallurgy at the Royal Institute of Technology in Stockholm.

Lee H. DeWald is metallurgical engineer for the National Cylinder Gas Co., manufacturers of industrial gases and welding equipment, Chicago. He receives mail at 828 Main St., Aurora, Ill.

Delmar W. Dow has the job of metallurgical engineer with the Permanente Metals Corp. at the Mead works, Spokane, Wash. His address in Spokane is 10014 Whittier St., Rt. 7.

Norman F. Dufty is now employed as technical assistant in the open hearth department of the Australian Iron and Steel Co., G.P.O. Wollongong, N. S. W. Australia.

Clarence W. Fabel is no longer employed by Remington Rand, Inc.; his present post is metallurgist with the Simonds Saw and Steel Co., Lockport, N. Y.

John H. Fuqua graduated from the Missouri School of Mines in January and has taken a job with the American Steel Foundries in the research laboratory at their Indiana Harbor plant. His address is 8020 S. Paulina St., Chicago.

William K. Heid is taking the supervisory training program at the Pontiac motor division of the General Motors Corp. He lives at 89 S. Roselawn Drive, Pontiac 20, Mich.

K. H. Heino is working at Noranda, Que., as technical assistant for Noranda Mines Ltd.

George O. Hiers works in the research laboratories of the National Lead Co., 105 York St., Brooklyn 1, N. Y.

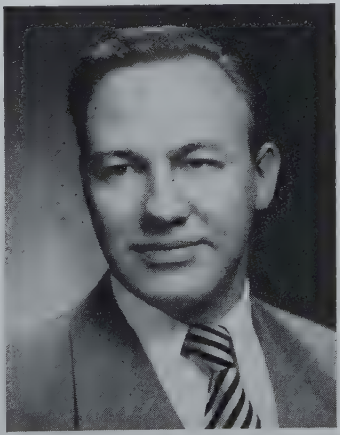
G. W. Higgins works for the Wehr Steel Co. at West Allis, Wis. His mail goes to 1743 S. 58th St., West Allis.

Erle J. Hubbard has resigned from his post as foundry superintendent in the piston ring division of the Koppers Co. and accepted the post of foundry manager of the Janney Cylinder Co., 7421 State Rd., Philadelphia, Pa.

Raymond A. Humphreys is metallurgical inspector and secretary of the materials review board of the Chase Aircraft Co., West Trenton, N. J.

William A. Johnson, formerly at Oak Ridge, Tenn., at the Clinton laboratory of the Monsanto Chemical Co., has gone to Pittsburgh, Pa., to work with the atomic power division of the Westinghouse Electric Corp.

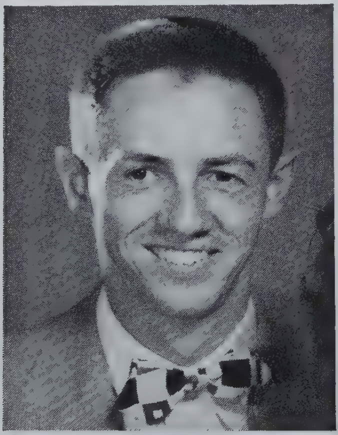
F. G. Kerry, who lives at 227 Glendale Rd., Scarsdale, N. Y., is manager of the engineering development department of Spaco Inc. (Air Liquide), oxygen producers and plant manufacturers, 529 Chrysler Bldg., New York City 17.


Charles H. T. Wilkins

Charles H. T. Wilkins, formerly metallurgist with the Copperweld Steel Co., heads the National-Standard Company's industrial fellowship at Mellon Institute in Pittsburgh. The Niles, Mich., company established the fellowship to investigate, in a long-range program, major problems related to the technology of fine wire products, such as tire bead wire, strand for heavy duty tire casings, music wire, and aircraft and textile wires. Also to be studied are problems related to flat strip for use in such products as piston rings, razor blades, clock and watch springs, pen points, and umbrella ribs. Mr. Wilkins was graduated from VMI and subsequently pursued advanced work in metallurgy at the University of Pittsburgh, receiving his M.S. degree there in 1946.

C. H. Lenhart has resigned as general superintendent of the steel plant of the Iron and Steel Division of the Kaiser Co., Fontana, Calif., and is replaced by **George B. McMeans.**

Arthur M. Monson has resigned from the Stearns-Roger Mfg. Co. to enter the realty business as salesman for Foster & Barnard, Inc., Chamber of Commerce Bldg., Denver.



William R. Opie

William R. Opie, a research assistant in metallurgy at MIT, received a degree of Doctor of Science in metallurgy in February. He is now employed as a research metallurgist by the Americal Smelting and Refining Co. in their central research department at Barber, N. J. His home address is 99 3rd St., Raritan Township, Fords, N. J.

• In Petroleum Circles

Paul T. Bail is working for the Bureau of Mines at Franklin, Pa. He used to be with the Stanolind Oil and Gas Co. in Wink, Texas.

Robert E. Bennett has joined the staff of the California Co. as a petroleum engineer at Harvey, La.

Paul Biggs resigned his job with the Stanolind Oil and Gas Co. at Big Lake, Texas, in February and is now employed by the Bureau of Mines at the petroleum and oil shale experiment station, Laramie, Wyo., as a petroleum engineer.

Guy Borden, Jr., assistant professor of petroleum engineering at Louisiana State University, has joined the production research department of the Stanolind Oil and Gas Co. at Tulsa, Okla.

Pete W. Cawthon, Jr., is completing a training program in the production department of the Phillips Petroleum Co. He finished work on a master of engineering degree, petroleum, at the University of Oklahoma last September. His degree will be granted in June.

Courtney E. Cleveland, after spending a number of years as geologist and exploration engineer with Bralorne Mines Ltd. of Vancouver, B. C., is now joining the geological staff of the Barnsdale Oil Co., 735 8th Ave. W., Calgary, Alta.

Robert R. Cocke, III, is a student at the University of Texas as well as part time engineer for the Railroad Commission of Texas. He lives in Apt. 1, 1807 Brazos, Austin, Texas.

Lynn R. Cook, Jr., can be reached at the Richmond Exploration Co., Apartado 93, Maracaibo, Venezuela.

William R. Corby is working for the Spartan Drilling Co. in the Golden Meadow field. At present he is doing work in both the production and drilling departments. His address is Box 506, Golden Meadow, La.

William J. Cox, as a petroleum geologist with the Creole Petroleum Corp., has gone to Barcelona, Anzoategui, Venezuela.

Jeral J. Crowder resigned as petroleum engineer from the Gulf Oil Corp. the first of the year to assume duties with the Quintana Petroleum Corp. as petroleum engineer for their producing and drilling department at Houston, Texas. Mr. Crowder attended Texas A&M College, studying petroleum engineering, and later he received a B.S. degree in mechanical engineering from the University of Houston.

Charles E. Davis, upon graduation from SMU in Dallas last May with a B. S. in geology, accepted a post with the Gulf Oil Corp. and was subsequently assigned to work in the N. Ward-Estes field of Ward and Winkler Counties, Texas, as a field geologist.

John O. Davis graduated from the University of Texas with a degree in petroleum engineering in February 1948 and since that time has been employed by the Jergins Oil Co. at Victoria, Texas, as production foreman.

Jack M. Denson graduated from the University of Texas last June with a B. S. degree in petroleum engineering and is now working for The Texas Co. as a petroleum engineer trainee. His address is Box 82, Buckeye, N. Mex.

William C. Dinger, formerly with the Houston Oil Co. of Texas, can be reached at Rock Springs, Wyo., where he is associated with the Mountain Fuel Supply Co.

John R. Evans is undergoing a one year training program as junior petroleum engineer for the Creole Petroleum Co., New York City, but receives mail in care of R. N. Dyer, Humble Oil and Refining Co., Houston, Texas.

Gearld I. Freeze, former instructor in the department of petroleum engineering at the University of Kansas, is now in the production department of the Phillips Petroleum Co., Box 1605, Hobbs, N. Mex.

Ionel I. Gardescu has been promoted to manager of the gas reserve department of the Transcontinental Gas Pipe Line Corp., Oil and Gas Bldg., Houston 2, Texas.

John T. Gary, with the Continental Oil Co., was transferred on Jan. 1 from district petroleum engineer at Wichita Falls, Texas, to production engineer in the general office at Ponca City, Okla.

Benjamin F. Hake can be reached at the geological department of the Gulf Oil Corp., Box 1166, Pittsburgh 30, Pa.

A. S. Hawley resigned in January from the Schlumberger Well Surveying Corp. and accepted the post of geologist with the Union Oil Co. in Bakersfield.

Jack P. Hays is production superintendent for the Intex Oil Co., Dallas, Texas, but his mail goes to 64 Bacon Bldg., Abilene, Texas.

Richard V. Hughes, formerly with the Penn Grade Crude Oil Assoc., has gone to Stanford, Calif., as geologist with the Stanford Research Institute.

Joe B. Justus graduated from Louisiana State and is working for the California Co., Ruston, La., as a production engineer.

James E. Keller has gone to Caracas, Venezuela, to work for the Creole Petroleum Corp. He is addressed there at Apartado 889.

Leslie A. Leuzinger is field engineer at Rangely, Colo., for the Union Pacific Railroad Co., oil development.

Robert L. Lindauer, Jr., receives mail at P. O. Box 357, Katy, Texas. He is a junior gas engineer with the Humble Oil and Refining Co.

Donald P. McCarthy, who was a student at the Montana School of Mines, is working in the gological department of the Sinclair Prairie Oil Co., Albuquerque, N. Mex.

John M. McLain is junior exploitation engineer with the Shell Oil Co., Houston, Texas, but gets his mail at 823 Main St., Hattiesburg, Miss.

Richard H. Zinszer can now be reached in care of the research department of the Union Oil Co. of California, Wilmington, Calif.

Obituaries

Charles E. LeNeve Arnold

An Appreciation by James L. Head

"Charlie Arnold died yesterday." Seldom does such sad news spread as rapidly as it did on Monday morning, Jan. 24, throughout the six floors of 25 Broadway, New York City, which house the vast New York organization of Anaconda. From top brass to office boy, from veteran to newest comer, all felt a distinct and unusual personal loss for he enjoyed the widest personal acquaintanceship in the entire organization. When the news of his passing was communicated to Inspiration, Potrerillos, Chuquicamata and Cananea, to each of which he had made definite contribution, to the Mining Club and to Institute Headquarters, it evoked the same reaction and added additional evidence of the particular esteem in which he was held by everyone who knew him. He was always the considerate, discerning, courteous, gentlemanly and unfailing friend of all. It is to such rare individuals that great organizations really owe that intangible thing, esprit de corps.

"None knew thee but to love thee,
Nor named thee but to praise"

For the record, Charles Everard LeNeve Arnold was born in Auckland, New Zealand, on Sept. 13, 1883, where he is survived by a father and brother. He attended Auckland University from March to November, 1901. His choice of a career in the mineral industry must have been an early one, for between 1900 and 1902 he was a bullion melter and assayer at Waihi, New Zealand. I do not know why he came to the States, but he did so in 1902, matriculating at the University of California, from which he was graduated in 1906. Here he was a schoolmate of Erle V. Daveler, James S. Wroth, John Tyssowski, and the late Spencer Browne. One of his favorite tales—and he was a raconteur par excellence—was of his first day in this country when he was initiated into American barbershop practice—and paid for the "works."

His summer vacations during college followed the traditional pattern of the embryo mining engineer of that day. We have such unglamorous items as machine hand, Union Iron Works, San Francisco; miner, Keystone Gold Mine, Amador City, Calif.; cyanide hand, North Star Corp., Grass Valley; sampler, Fresno Copper Corp., Gordon, Calif.; and draftsman at Ely for Nevada Consolidated. Following graduation, he was from 1905 to 1911 construction and mine engineer for the Girous Consolidated Mines Co. at Kimberley, Nev., and I think it was here that he met the future Mrs. Arnold, Miss Genevieve Mills.

Moving on to Inspiration, the newest addition to the porphyry copper group, he was mine engineer there from 1911 to 1918 and had much to do with the sound planning which was the basis for the success of that operation. In 1918, he was called to New York City and it is not surprising that with this background and wealth of experience he was designated to function in a similar capacity in the planning and development of Anaconda's newest venture, the great Andes Copper Mining Company's mine at Potrerillos, Chile. In 1924 he became assistant to William Wraith, vice-president, Andes Copper Mining Co., Greene Cananea Copper Co., and Inspiration Consolidated Copper Co., and he served in this capacity until Mr. Wraith's retirement at the end of 1945. Subsequently he continued in close touch with the operations of these companies.

All of which adds up to the fact that Arnold was an exceedingly capable engineer. So useful, dependable, and conscientious was he in the ever essential "information please" capacity, that he was never quite able to shake off the shackles and inhibitions which so often prevent a fine engineer from achieving top executive rank.

He became a member of AIME in 1906 and was one of headquarters' dependables. He was particularly active in the New York Section, having served as secretary, 1934, treasurer 1935-37, vice-chairman, 1938 and chairman 1939. Always intimately connected with the Annual Meetings of the Institute in New York City, his designation as chairman of any committee, annual banquet, informal dance, reception, etc., insured the success of that particular function. He had looked forward to attending the recent Annual Meeting in San Francisco. At the time of the organization of the Mining Club in New York City in 1930, he was invited to become a charter member and he was elected to membership in the Mining and Metallurgical Society of America in 1939.

The fact that he had suffered an initial heart attack on Labor Day, 1940, did not deter him from carrying on, and although he was possibly living on borrowed time, he continued to be the genial and unforgettable Charlie Arnold.

D. B. McAllister

All who have regularly attended the Annual and Regional meetings of the Institute will be saddened to hear of the death of one of those whose presence has contributed so much to those occasions—D. B. McAllister, of San Luis Potosi, Mexico. "Mac," as he was known to so many friends, and to his wife as well, died on March 28. The last time we saw him was at the stag dinner at the San Francisco meeting, when we sat next to him. At that time he seemed a bit repressed for Mac, but we had no idea this might have been a foreboding. We had exchanged pleasantries at many other meetings, and he was always making a joke of something, or assuming a character that he thought might shock strangers. We are sure that a host of Institute members must have had a cordial invitation to visit at the McAllister home in San Luis Potosi, as we have, and no doubt many have been able to drop off there, as we have not. His good wife was always with him at meetings, endeavoring to keep him straight, as he explained. She will be an especially lonely woman. The Institute has lost not only a competent engineer but an active member and one who, although not on many of its technical programs as an author, helped greatly to make attendance at its meeting a satisfying experience.—*E. H. R.*

Daniel Bruno McAllister (Member 1920), was a native Bostonian, born in 1881, and a Harvard graduate, class of 1905. The year after graduation, he worked as a surveyor for the Pennsylvania tunnels in New York City, and for the next three years held engineering positions in West Virginia, Michigan and New York. In 1910 he became an engineer for the Cia. Metalurgical Mexicana, beginning a lifelong association with Mexican mining which was interrupted only by four years in New Mexico and two years with New York State's Public Service Commission. In 1925 he leased the El Barreno mine in San Luis Potosi, and for the past twenty-three years had resided in that city, actively engaged in mining enterprises.

Necrology

Date Elected	*Name*	*Date of Death*
1939	LeGrande D. Barrett	Unknown
1920	Lucius A. Butler	Jan. 23, 1949
1946	Herschel E. Denton	March 30, 1949
1931	Cecil Drake	Jan. 31, 1949
1917	Frederic W. James	Dec. 28, 1948
1920	Daniel B. McAllister	March 28, 1949
1915	Alexander G. McGregor	March 4, 1949
1936	Arthur S. Nichols	Feb. 18, 1949
1916	Chase S. Osborn	April 11, 1949
1948	Thomas M. Robertson	Aug. 16, 1948
1940	William A. Rose	Jan. 30, 1949
1904	David R. Thomas	March 4, 1949

HAROLD E. WILLSON (Member 1916), West Virginia mining engineer who headed his own firm at Oak Hill, is dead. A graduate of the University of Illinois, class of 1916, Mr. Willson worked with various West Virginia coal companies before forming his own engineering organization, the H. E. Willson Engineering Co., in 1918, which he operated until the time of his death.

Proposed for Membership

Total AIME membership on March 31, 1949, was 15,363; in addition 4082 Student Associates were enrolled.

ADMISSIONS COMMITTEE

James L. Head. Chairman: Albert J. Phillips, Vice-Chairman: George B. Corless, T. B. Counselman. Ivan A. Given, George C. Heikes, Richard D. Mollison, and Philip D. Wilson.

Institute members are urged to review this list as soon as the issue is received and immediately to wire the Secretary's office, night message collect, if objection is offered to the admission of any applicant. Details of the objection should follow by air mail. The Institute desires to extend its privileges to every person to whom it can be of service but does not desire to admit persons unless they are qualified.

In the following list C/S means change of status: R, reinstatement; M, Member; J, Junior Member; AM, Associate Member; S, Student Associate; F, Junior Foreign Affiliate.

ALABAMA

Bessemer — SAUNDERS, ARCHIE OLIVER. (AM). Service engineer, Independent Pneumatic Tool Co.

Florence—SHERWIN, RALPH SIDNEY. (R—M). Vice-president in charge of alumina development, Reynolds Metals Co.

ARIZONA

Hayden—TUCK, FRANK JOSEPH. (R—M). Mill superintendent, Kennecott Copper Corp.

Jerome—FULTON, DAVID HOLMES. (C/S—S-J). Stope engineer, Phelps Dodge Corp.

Superior — ANDERSON, RICHARD COLEMAN. (C/S—J-M). Mining engineer, Magma Copper Co. ROBERTS, ALVAN E., JR. (C/S—S-J). Transitman, engineering dept., Magma Copper Co.

ARKANSAS

El Dorado — McDONALD, PAUL CRANFORD. (R,C/S—J-M). Field superintendent, C. H. Murphy & Co.

Little Rock—HESS, ALBERT JOHN. (M). Mining engineer, The Milwhite Co.

CALIFORNIA

Alhambra—NASH, WILLIAM FRANCIS, JR. (M). Staff consultant, C. F. Braun & Co.

Berkeley—PIETROKOWSKY, PAUL. (C/S—S-J). Research metallurgist, University of California. SCHUELER, DORNER TURNER, JR. (C/S—S-J). Chief engineer, Industrial Minerals & Chemical Co.

Buena Park—WRIGHT, CHARLES CATHBERT. (J). Chemist, Oil Well Research, Inc.

Inglewood—GRAVES, DOYLE THEODORE. (J). Subsurface engineer, Union Oil Co. of California.

Long Beach—VAN RIPER, JOHN. (M). Petroleum engineer, Continental Oil Co.

Los Angeles—BOYES, RALPH LEE. (J). Assistant field engineer, Union Oil Co. EVANS, THOMAS O. (M). Mining engineer, Atchison, Topeka & Santa Fe Railway Co. JAKOSKY, JOHN JAY, JR. (C/S—S-J). Engineer, International Geophysics, Inc. MARSHALL, IRA ALLISON, JR. (J). Production equipment engineer, Shell Oil Co.

Ojai — DOIG, KEITH. (J). Junior mechanical engineer, Shell Oil Co.

San Francisco — MORBIO, CARLO SUTRO. (AM). Consultant.

South Gate — LIETZOW, CLYDE HARVEY. (J). Sales engineer, Pelton Water Wheel Co.

South Pasadena—PAUL, WILLIAM GLAE, JR. (J). Representative, Kobe, Inc.

Ventura—WEST, W. FRANK. (C/S—S-J). Petroleum engineer-trainee, Continental Oil Co.

Whittier—FAULL, RICHARD FRANCIS. (M). Manager, La Habra Laboratory, California Research Corp.

Wilmington—FEI, ROBERT FLEMING. (C/S—S-J). Petroleum engineer-trainee, Union Oil Co.

COLORADO

Denver—BOARD, VIRGIL LEE. (M). President, Western Exploration & Management Co. SNEDDEN, HENRY DAVID. (R,C/S—J-M). Metallurgist, Humphreys Investment Co.

Golden — LOEB, MICHEL JEAN. (C/S—S-J). Graduate student, Colorado School of Mines.

Trinidad—DUREN, JACK DANIEL. (C/S—S-J). Junior exploitation engineer, Shell Oil Co.

CONNECTICUT

Manchester—DAY, ROBERT B. (C/S—S-J). Metallurgical engineer, Linde Air Products Co.

New Haven—HIBBARD, WALTER ROLLO, JR. (C/S—J-M). Assistant professor of metallurgy, Yale University.

Ridgefield — MARTIN, MAURICE HENRI. (M). Chief engineer, Schlumberger Well Surveying Corp.

Stratford — CLARK, HOWARD T., JR. (M). Research metallurgist, Remington Arms Co.

Waterbury — DUBPERNELL, GEORGE. (M). Manager, Waterbury laboratories, United Chromium.

DISTRICT OF COLUMBIA

Washington—WARNCKE, RUSSELL GEHRING. (R,C/S—JA-M). Mining engineer, safety branch, Bureau of Mines. WRIGHT, WILLIAM EDWARD. (C/S—S-J). Commodity specialist (ceramics), U. S. Tariff Commission.

GEORGIA

Marietta—WILLINGHAM, CHARLES BERRIEN. (M). Sales engineer, Joy Mfg. Co.

IDAHO

Kellogg — McDONALD, JAMES THOMAS. (C/S—S-J). Surveyor, Bunker Hill & Sullivan Mining & Concentrating Co.

Wallace — NELSON, VERNON JEROME. (AM). Division manager, Coeur d'Alene mining division, Washington Water Power Co.

ILLINOIS

Chicago — DICKTER, IRVING A. (M). Plant metallurgist and chemist, Goodman Mfg. Co. DYER, HARRY BISHOP. (R,C/S—S-J). Research metallurgist, Republic Steel Corp. LEWY, JAMES KAY. (J). Metallurgical trainee, Inland Steel Co. SILVERSTEIN MITCHELL. (C/S—S-J). Metallurgist, Silverstein & Pinsof.

Marshall — NAERT, GILBERT ALPHONSE. (J). Petroleum engineer, Ohio Oil Co.

Mt. Vernon—SHICK, ROBERT L. (R,C/S—S-M). District petroleum engineer, Sohio Petroleum Co.

Olney — STORSETH, BILLY JOE. (J). Sales engineer, Lane-Wells Co.

INDIANA

Evansville—VAUGHAN, JERRY R., JR. (R,C/S—S-M). Petroleum engineer, Sun Oil Co.

West Lafayette—GUARD, RAY W. (C/S—S-J). Instructor, metallurgical engineering, Purdue University.

KANSAS

Eureka—McGINNIS, WARD ALLYN. (M). President and general manager, Ring-Mac Oil Co.

Madison—JONES, FRANK T. (J). Junior engineer, Cities Service Oil Co.

Manhattan — WIKLE, KEITH GORDON. (R,C/S—S-J). Professor of engineering metallurgy, Kansas State College.

Wichita — HARWOOD, WARREN KENNETH. (R,C/S—S-J). Secretary-treasurer, Harwood Oil Co.

KENTUCKY

Wheelwright—ANDERSON, ROBERT BRADFORD. (C/S—S-J). Engineer, Inland Steel Co.

LOUISIANA

Baton Rouge—PYBURN, THOMAS DENNIS. (C/S—S-J). Service engineer, Baroid Sales Div., National Lead Co.

New Orleans—WATSON, MARTIN C. (M). Assistant area manager, Southeastern area, Schlumberger Well Surveying Corp.

Shreveport—LOGAN, GEORGE LYONS. (C/S—S-M). Chief petroleum engineer, Lyons & Prentiss.

MARYLAND

Berwyn Heights — BROWNING, JAMES SCOTT. (C/S—S-J). Metallurgist, Bureau of Mines.

MASSACHUSETTS

Lexington — HURLEY, PATRICK MASON. (C/S—J-M). Assistant professor, MIT.

Pittsfield — GARDNER, FRANK STREETER. (C/S—J-M). Metallurgist, General Electric Co.

Springfield—NAST, PAUL H. (M). Mining drill sales supervisor, Worthington Pump & Machinery Corp.

MICHIGAN

Detroit—SCHMIDT, WILLIAM ATKINSON. (C/S—S-J). Graduate trainee, Cadillac Motor Car Div., General Motors Corp.

MISSISSIPPI

Clinton — MATHER, KATHARINE KNISKERN. (C/S—J-M). Head, petrography section, concrete research div., Waterways Experiment Station.

Hattiesburg — McLAIN, JOHN McAFEE. (C/S—S-J). Junior exploitation engineer, Shell Oil Co.

West Jackson—JEFFREYS, GEOFFREY. (R-M). Consultant, petroleum geology.

MISSOURI

Herculaneum — MICKLE, RALPH GLOVER, JR. (R,C/S—S-J). Metallurgist, St. Joseph Lead Co.

St. Louis—VOGT, JOSEPH CHARLES. (C/S—S-J). Department head, Laclede Gas-Light Co. SCHRAUT, JOSEPH ANTHONY, JR. (C/S—S-J). Geologist, concrete laboratory, Aggregate Investigation, U. S. Army Engineers.

Webster Groves—CHUBB, WALSTON. (C/S—S-J). Ludlow Saylor research fellow, Missouri School of Mines & Metallurgy.

MONTANA

East Helena—MORAN, FRANK C. (M). Assistant superintendent, East Helena plant, American Smelting & Refining Co.

NEW JERSEY

Riverside — TURSI, PETER BUDD. (M). Metallurgist, Riverside Metal Co.

South Orange—COOPER, HERBERT JAY. (C/S—S-J). Engineer of research & development, Cooper Alloy Foundry Co.

NEW YORK

Brooklyn — DORKOVICH, VICTOR. (R-M). Senior member, R. S. Aries & Associates. KANE, NATHAN MILTON. (R,C/S—S-J). Piping engineer, Lummos Co.

Mt. Vernon — DONELLON, JOHN BERNARD. (M). Process control chemical engineer, Callite-Tungsten Corp.

Natural Bridge—BLAZOVIC, PHIL, JR. (C/S—J-M). Plant metallurgist, Wheeling Steel Corp.

New York—CULBERTSON, FRANCIS DORAM. (J). Sales engineer, Ingersoll-Rand Co. TUTHILL, MARSHALL WALLACE. (AM). President, Tuthill & Co.

Niagara Falls—DEMMON, NEWTON KEITH. (M). Service engineer, National Carbon Co.

Schenectady—CECH, ROBERT EDWARD. (C/S—S-J). Research assistant, physical metallurgy section, General Electric research laboratory. MARTIN, DONALD LUTHER. (C/S—J-M). Research metallurgist, General Electric Co.

Sherrill — MOULTON, WALTER EVERSON. (M). Chief metallurgist, Oneida, Ltd.

Upton — TEITEL, ROBERT JERRELL. (C/S—S-J). Associate metallurgist, Brookhaven National Laboratories.

OHIO

Cleveland — HAMJIAN, HARRY JOHN. (J). Aeronautical research scientist, National Advisory Committee for Aeronautics.

Columbus — BRISON, ROBERT JOSHUA. (R,C/S—S-J). Research engineer, Battelle Memorial Institute. SCHIPPEREIT, GEORGE HENRY. (C/S—S-J). Research engineer, Battelle Memorial Institute.

Middletown — ROTHE, LEONARD WILLIAM. (C/S—S-J). Metallurgical trainee, Armco Steel Corp.

OKLAHOMA

Konawa — QUILLIN, JOHN PAUL. (C/S—S-J). Trainee engineer, Standard Oil of Ohio.

Tulsa—HALEY, KENNETH WENDELL. (M). Chief petroleum engineer, Deep Rock Oil Corp. HUZAREVICH, JULIAN ELI. (C/S—S-J). Junior petroleum engineer, Stanolind Oil & Gas Co. PERRY, RALPH. (M). Director, Core laboratory, Gulf Oil Corp. WILSON, SAM A. (M). Chief engineer, Maloney-Crawford Tank and Mfg. Co.

PENNSYLVANIA

California — HESS, WILLIAM EDWARD. (M). General superintendent, Vesta Shannopin coal division, Jones & Laughlin Steel Corp.

Commodore—SIMPSON, JOHN FAIRFULL, JR. (C/S—S-J). Junior mining engineer, Clearfield Bituminous Coal Corp.

Frackville — ROSELLA, JOHN JERRY. (M). Mining engineer, Bureau of Mines.

Palmerton—RAMSEY, PAUL WILLARD. (C/S—J-M). Metallurgist, development engineering division, New Jersey Zinc Co. (of Pa.).

Philadelphia—MULDAWER, LEONARD. (J). Assistant professor of physics, Temple University.

Pittsburgh — ELKIN, JAMES ELWELL. (M). General superintendent, Coal department, Duquesne Light Co. HENSCHEN, ARTHUR EVANS. (C/S—S-J). Sales engineer, Aluminum Co. of America. VOELKER, HENRY RONALD. (C/S—S-J). Student-graduate, doing nondegree studies at the University of Pittsburgh.

State College — NIELSEN, RALPH FREDERICK. (M). Assistant professor of petroleum and natural gas engineering, Penn. State College. SCHANZ, JOHN J., JR. (C/S—S-J). Instructor, mineral economics, Penn. State College.

Summit Hill—KUEBLER, CHARLES SWANK. (R,C/S—J-M). District superintendent, Lehigh Navigation Coal Co.

SOUTH DAKOTA

Vermillion—ROTHROCK, E. P. (M). State geologist, South Dakota Geological Survey.

TEXAS

Arp—ROGERS, WILLIAM JEROME. (R,C/S—S-M). District engineer, Sinclair Prairie Oil Co.

Austin—McKETTA, JOHN J., JR. (M). Associate professor of chemical engineering, Univ. of Texas.

Beeville — McDOWELL, OLIVER GEORGE. (J). District engineer, Houston Oil Co. of Texas.

Dallas—BRANSON, UZZELL S., JR. (M). Reservoir analyst, James A. Lewis Engineering. CHAMBERS, ELLEN ADELE. (J). Junior chemist, Atlantic Refining Co. LORING, WILLIAM THOMAS. (C/S—J-M). Petroleum engineer, Core Laboratories, Inc. MARSHALL, DAN LANDRY. (C/S—J-M). Geological and petroleum engineer, DeGolyer & MacNaughton.

El Paso—BARSA, DAVID J. (J). Fire assayer, Phelps Dodge Refining Corp. HERPERS, FRANK MANUEL. (M). Assistant general foreman, casting department, Phelps Dodge Refining Corp. LAUREL, LOUIS PHILLIP. (AM). Manager, machinery department, Mine & Smelter Supply Co. McNAMARA, FRANCIS ROBERT. (M). District manager, Gardner Denver Co. RINDOM, VIRGIL JENNINGS. (M). Chemist, Phelps Dodge Refining Co.

Fort Worth — FORBES, HENRY FRASER. (R,C/S—J-M). Region petroleum engineer, Continental Oil Co. WILSON, JOHN HUMAN, II. (C/S—S-J). Pumper, Fred H. Manning, Inc.

Houston—AUSTIN, JESS WILLIAM. (C/S—S-J). Junior petroleum engineer, Stanolind Oil & Gas Co. DUPONT, HARRY ALVIN. (R,C/S—S-J). Petroleum engineer, Phillips Petroleum Co. GETZENDANER, AVELETTE EDMOND. (M). Zone geologist, Gulf Products Division, Gulf Oil Corp. LLOYD, JOHN MORGAN, JR. (J). Petroleum engineer, Stanolind Oil & Gas Co. RIDLEY, ROBERT P. (R,C/S—S-M). Reservoir engineer, The Texas Co. WAECHTER, JACK WILLIAM. (C/S—S-M). Petroleum production engineer, California Co. WATSON, JOSEPH PRESLEY, JR. (J). District engineer, Cities Service Oil Co.

Katy—LINDAUER, ROBERT LEE, JR. (C/S—S-J). Junior gas engineer, Humble Oil & Refining Co. SHARP, CARLE FARRELL, JR. (C/S—S-J). Junior reservoir engineer, Humble Oil & Refining Co.

Kilgore — ELSEY, HARRY NEILL. (M). District machinery sales engineer, Republic Supply Co.

Lubbock — WHITIS, EMETT D. (C/S—S-AM). Junior petroleum engineer, Stanolind Lubbock district officer, reservoir section, Stanolind Oil & Gas Co.

Midland — BANISTER, RICHARD CARL. (C/S—S-J). Geologist, Standard Oil Co. of Texas. WEAVER, BILLY RAY. (C/S—S-J). Associate engineer, Tide Water Associated Oil Co.

Odessa — SHANOR, PAUL JAMES. (C/S—S-J). Engineer, Phillips Petroleum Co. SHILCUTT, CHARLES E. (AM). District superintendent, Lane-Wells Co.

Premont—COIL, CARL LACY. (J). Junior petroleum engineer, Sun Oil Co.

UTAH

Clearfield—FUNK, ROBERT LOUIS. (R,C/S — S-M). Sales representative, Timken Roller Bearing Co.

Lark—BOYD BENTON. (R,C/S—J-M). Mine superintendent, Lark mine, U. S. Smelting Refining & Mining Co. RUBRIGHT, RICHARD DICKSON. (M). Associate geologist, U. S. Smelting Refining & Mining Co.

Midvale—FRESH, NEIL. (M). General mine foreman, U. S. Smelting Refining & Mining Co.

Price — REEVES, JOHN ALLEN. (C/S—S-J). Assistant engineer, Independent Coal & Coke Co.

Salt Lake City—MATES, ROBERT RICHARD. (C/S—S-J). Metallurgist, Bureau of Mines.

WASHINGTON

Spokane—OMELKA, LEON VLADIMIR. (C/S—S-J). Testing engineer, Permanente Metals Corp.

WEST VIRGINIA

Cedar Grove — PALLER, CARL ADOLPH. (M). Mining engineer, Warner Collieries Co.

Charleston—BRIBER, ALEX AUSTIN. (C/S—S-J). Junior engineer, Commercial Testing & Engineering Co.

Fairmont — SCHIMMEL, JOHN THRALL. (C/S—S-J). Time study engineer, Consolidation Coal Co.

Weirton — HARRIS, RALPH WILLIAM. (C/S—S-J). Tester, first class, Weirton Steel Co.

Wheeling — FLANDERS, ALBERT CHARLES. (R,C/S—S-M). Mine superintendent, Carbola Chemical Co.

ALASKA

Anchorage — McANERNEY, JOHN MARSHALL. (R,C/S—J-M). Chief geologist, Alaska district, Corps of Engineers.

Juneau — WARFIELD, ROBERT STEWART. (C/S—S-J). Assistant mining engineer, exploration and development division, Bureau of Mines.

BRITISH COLUMBIA

Victoria—LINEHAM, JOHN DONALD. (C/S—S-J). Assistant controller, petroleum and natural gas, Government of Province of British Columbia.

ONTARIO

Falconbridge — WARNER, JOHN PELHAM. (J). Graduate trainee, Falconbridge Mines, Ltd.

Missanabie—CARROLL, JAMES WILLARD. (M). Mine superintendent, Renabie Mine.

Ottawa — KOZLOWSKI, HENRY JULIAN. (R,C/S—S-J). Research metallurgist, Bureau of Mines, Ottawa.

QUEBEC

Thetford Mines—JENKINS, GEORGE FRENCH. (R,C/S—JA-M). General manager, Asbestos Corp.

NEWFOUNDLAND

Buchans — SWANSON, ERIC ALFRED. (M). Assistant geologist, Buchans Mining Co.

CUBA

Pinar del Rio—HENDERSON, JAMES GEORGE. (M). Engineer in charge, Minas de Matahambre S. A.

BOLIVIA

Camiri—ROMERO LOZA, ALFONSO. (M). Production superintendent, Yacimientos Petroliferos Fiscales Bolivianos, Camiri Oilfield.

VENEZUELA

Barcelona—COX, WILLIAM JONES. (C/S—S-J). Petroleum geologist, Creole Petroleum Corp. WICKENHAUSER, LEONARD JOHN. (M). Reservoir engineer, Mene Grande Oil Co.

BELGIUM

Montigny le Tilleul—RICHARD ROBERT VALERE. (M). Director general de Usines et Acieries Allard (Belgium).

ENGLAND

London—DOWNWARD, BRIAN SUTTON. (M). Geologist, Whitehall Securities Corp. VOCE, ERIC. (M). Chief metallurgist, Copper Development Association.

ITALY

Vado Ligure — CODA, AGOSTINO GIACOMO. (M). Manager, Vado zinc works, Soc. di Montepoui.

BELGIAN CONGO

Elisabethville — VAN VINCKEROY, VICTOR. (R-M). Assistant to consulting engineer, Union Minière du Haut Katanga.

EGYPT

Cairo — TRAFFORD, EDWARD LE MARCHANT. (M). Chief exploitation engineer and manager, Anglo-Egyptian Oilfields, Ltd., Shell "Group."

SAUDI ARABIA

Dhahran—FLITTIE, CLIFFORD GILLILAND. (C/S—S-J). Assistant geologist, Arabian American Oil Co.

INDIA

Bihar — RAVEN, JOHN MANUEL. (M). Mine agent, Indian Copper Corp.

Champion Reefs—KERNICK, JOHN EDDY, JR. (M). Assistant treatment officer, Champion Reef Gold Mines of India.

NEW SOUTH WALES

Ashfield — KING, MALCOLM GEORGE. (AM). Research engineer, Australian Blue Asbestos, Ltd.

Cobar — BEATTIE, GEORGE ANDREW. (AM). Assistant underground superintendent, New Occidental Gold Mines.

Student Associates

Elected March 16, 1949

Louis L. Bergan	Mont. Sch. Mines
John B. Clark	Carnegie Inst. Tech.
Henry L. Hurley	S. Dak. Sch. Mines
Nelson Severinghaus, Jr.	Univ. Ariz.
Ernest Stewart (Rein.)	Univ. Ariz.

Elected April 20, 1949

Alfred Ainsworth, Jr.	Univ Wichita
Harold G. Alford	Univ. Tulsa
Carlton D. Bailey, Jr.	Mich. Min. & Tech.
Emmet H. Balch, Jr.	Univ. Tulsa
Laurence H. Beal	Univ. Calif.
John T. Berry (Rein.)	Texas A&M
Christy R. Bickford, Jr.	Univ. Wichita
George A. Bivens	Univ. Mich.
Dewey A. Bjork	Mich. Min. & Tech.
John L. Black	Otago Univ.
Robert D. Black	Fenn Coll.
Edwin A. Blackey, Jr.	Univ. N. H.
Carter H. Boswell	Univ. S. Calif.
Maurice H. Brady	Univ. Nev.
Leo R. Brammer, Jr.	Univ. Tulsa
Stanley G. Britton	Univ. Tulsa
Robert G. Broady	Univ. Kans.
Douglas E. Brown	Colo. Sch. Mines
Frank L. Brown	Univ. Tulsa
Harry L. Brown	Colo. Sch. Mines
George W. Bruce	Univ. Wichita
Robert Bulla	Mo. Mines & Met.
John Bychok	Univ. N. H.
David T. Byrne	Univ. Pittsburgh
Francis W. Byrne, Jr.	Univ. Pittsburgh
John P. Campbell, Jr.	La. State Univ.
John D. Carmichael	Univ. S. Calif.
Robert F. Cassel	Univ. Tulsa
Richard F. Chew	Univ. S. Calif.
Alfred J. Christenson	Mich. Min. & Tech.
Jesse L. Clifton	Univ. Okla.
Otis H. Collingwood	Univ. Wichita
William W. Collins	Mo. Mines & Met.
Christian M. M. Colson	Mo. Mines & Met.
George P. Conard, 2nd	MIT
James P. Corbitt	Univ. Tulsa
James A. Crate	Univ. Pittsburgh
Joseph D. Crites	Mo. Mines & Met.
J. W. Cunningham	Carnegie Inst. Tech.
Robert B. Daniel	Colo. Sch. Mines
Richard M. Dannenberg, III	Univ. Okla.
William J. Davies	Wash. Univ.
Charles DeAinza	Univ. S. Calif.
Robert W. Decker	MIT
Earle V. Dempsey	Univ. Nev.
Joseph G. Dempsey	Univ. Tulsa
Willis L. Dotson	Colo. Sch. Mines
Thomas H. Duff	Univ. Okla.
Clifford W. Dunham	Univ. S. Calif.
James P. Dunn	Univ. Tulsa
Jacob Eichler	CCNY
Myron R. Elliott	Univ. Calif.
William D. Elliott	Univ. Tulsa
John C. Ellis	La. State Univ.
Philip L. Essley, Jr.	Univ. Tulsa
Irwin W. Evans	Ill. Inst. Tech.
Johnie W. Falls	Univ. Tulsa
Freeman K. Farman	Univ. N. H.
Anthony M. Federico	Fenn Coll.
Richard K. Fisher	Penn. State
Vernon M. Flemming	Univ. Tulsa
Ludwig J. Frank	Univ. Texas
David J. W. Frazer	Colo. Sch. Mines
Walter L. Gage	Fenn Coll.
Harvey M. Gandy	La. State Univ.
William W. Gardner	Texas A&M
Domenick J. Garibotti	Fenn Coll.
D. Gogai	Univ. S. Calif.
Stanley Goldstein	Univ. Wichita
John J. Gouty	St. Louis Univ.
Robert Gray	Univ. Tulsa
Arnold W. Greenius	Univ. B. C.
Charles A. Gyorgak	Fenn Coll.
William C. Hagel	Cornell Univ.
Hugh E. Hanagan	Univ. Wichita
Kenneth E. Hard	Univ. S. Calif.
John W. Harris, Jr.	Univ. Okla.
William E. Harris	Univ. Texas
David Y. Heiman	Univ. Ill.
Jack H. Herring	Univ. Texas
Justin L. Holte	Univ. S. Calif.
Charles C. Horace	Univ. S. Calif.
Harry L. Horton	Texas A&M
Don L. Huber	Univ. Wichita
David D. Hughes	St. Louis Inst. Tech.
Earl E. Jackson	Mo. Mines & Met.
Herbert D. Jirrels	Univ. Wichita
H. M. Johannesmeyer	Mo. Mines & Met.
Horace A. Johnson	Case Inst. Tech.
Richard W. Johnson	Univ. Tulsa
Lewis J. Jones	Fenn Coll.
Thomas I. Jones	Colo. Sch. Mines
David C. Jonson	Colo. Sch. Mines
Sylvester C. Judge, III	St. Louis Univ.
Norman Karlin	CCNY
Byron L. Keil	Mo. Mines & Met.
James B. Kelly, Jr.	Texas Mines & Met.
Isabelle A. Keroack	Univ. N. H.
Kon-Hock Khaw	Univ. Calif.
Richard P. Kirkpatrick	Texas A&M
Max L. Knotts	Univ. Tulsa
Nathan R. Koenig	Texas Mines & Met.
Arthur M. Krause	Mo. Mines & Met.
Eddie B. Krider	Univ. Tulsa
William J. Krummel, Jr.	St. Louis Univ.
S. Victoria Krusiewski	Univ. N. H.
Melvin E. Landry	La. State Univ.
Billy J. Langford	Texas A&M
Hans G. Lanner	Fenn Coll.
Edward C. Leonard	Mo. Mines & Met.
Godfrey R. Lindsay	CCNY
Shu-I Liu	Case Inst. Tech.
Donald D. Lloyd	Univ. S. Calif.
George R. Locker	Univ. Tulsa
Arthur H. Lutts	Mont. Sch. Mines
Eugene S. Machlin	MIT
Arthur F. Malicoat	Texas A&M
Carroll F. Malinowsky	Univ. Kans.
Roy E. Margenau	Mich. Min. & Tech.
Cornelius J. Mastright	Univ. S. Calif.
Shannon L. Matheny	Univ. Calif.
Walter B. Mazgaj	Fenn Coll.
Gerald W. McCarty	Univ. Wichita
John McClurg	Fenn Coll.
Robert E. McDonald	Mo. Mines & Met.
Ronald B. McGregor	Univ. Melbourne
Malcolm C. McPhail	Texas A&M
Ira J. Meador	Univ. Kans.
Daniel F. Merriam	Univ. Kans.
Robert N. Miller	Univ. Texas
Edgar N. Mills	Univ. Tulsa
John A. Misencik	Fenn Coll.
Stuart Montgomery	Colo. Sch. Mines
Albert H. Moon	Cornell Univ.
David W. Morgan	Univ. B. C.
Francis D. Nelson	Ill. Inst. Tech.
Roland E. Nichols	Univ. Okla.
William R. Nicholson	Univ. Melbourne
Robert E. Nolan, Jr.	Lehigh Univ.
Charles L. Norton, Jr.	Penn. State
William J. O'Brien, Jr.	Univ. Texas
Alexander E. O'Meara	Otago Univ.
John A. O'Rourke	Colo. Sch. Mines
Robert L. O'Shields	Mo. Mines & Met.
Charles A. Peek	Mo. Mines & Met.
George Pepperdine	Univ. Calif.
George K. Peters	Univ. Pittsburgh
Robert L. Pierce	Univ. S. Calif.
Joseph Pinto	Penn. State
Flave Pledger	Univ. Penn.
Howard Porper	CCNY
Alan W. Postlethwaite	MIT
Daniel J. Pressler	Fenn Coll.
Vance Price	W. Va. Univ.
Edward E. Pugh, Jr.	Univ. Okla.
John G. Quetsch, Jr.	Fenn Coll.
Habib S. Rahme	Univ. Okla.
Paul K. Randall, Jr.	Hamilton Coll.
J. K. Rauschenberger	Univ. Pittsburgh
Richard B. Reynaud	La. State Univ.
Wilbur E. Richey	Texas Mines & Met.
Michael S. Rodolakis	Mo. Mines & Met.
Alvin E. Roemershauser	La. State Univ.
Gerald F. Rome	La. State Univ.
Jack H. Ronsick	Univ. Kans.
J. Charles St. Cyr	Univ. N. H.
Henry A. Sakowski	Univ. N. H.
Orhan M. Salman	Univ. S. Calif.
Warren C. Sander	Univ. N. H.
Raymond J. Schneider	Univ. Okla.
Richard R. Schuppert	Univ. S. Calif.
Robert E. Shafer (Rein.)	Univ. Alaska
Seymour L. Sharps	Univ. N. H.
George R. Shaw	Univ. Tulsa
Gerald E. Sherrod	Univ. Tulsa
Matthew J. Sladic	Univ. Pittsburgh
Milan Slafkovsky	Univ. Mines, Ostrava
Robert C. Slusher	Mo. Mines & Met.
Robert J. Smith	Fenn Coll.
Robert L. Smith	La. State Univ.
James W. Snarr	Univ. Ill.
Thomas K. Snodgrass	Univ. Tulsa
Milton A. Sobie	Mo. Mines & Met.
Michael Stasko	Fenn Coll.
David S. Stear	Univ. Tulsa
H. J. Steinbuchel, Jr.	Colo. Sch. Mines
Peter R. Stevens	Univ. N. H.
Jerome Stone	CCNY
Jack F. Thompson	Fenn Coll.
Charles V. Thrash	Fenn Coll.
Morris S. Todd	Univ. Okla.
Henry H. Tonking, Jr.	Mo. Mines & Met.
Charles E. B. Tothill	Mo. Mines & Met.
William J. Trahar	Univ. Melbourne
William J. Turner	Univ. Tulsa
Lawrence H. Van Vlack	Univ. Chicago
William R. Varner	Mo. Mines & Met.
Alberto M. Vazquez	Univ. Calif.
Rollin C. Vickers	Mo. Mines & Met.
Lee A. Ward	Univ. S. Calif.
James T. Watson, Jr.	Univ. Texas
Thomas H. Weidman	Mo. Mines & Met.
Adolf Weigand	CCNY
Harold A. Weinland	Mo. Mines & Met.
Leslie J. Whaley	La. State Univ.
Paul L. Widener	Cornell Univ.
Thomas Winterfeld	CCNY
Carl E. Wojan	Mo. Mines & Met.
David S. Wolinsky	CCNY
Walter G. Wong	Univ. Calif.
Alfred C. Wright	Univ. S. Calif.
Charles E. R. Wuestner	Wash. Univ.
Robert A. Wycoff	Univ. S. Calif.
Clifford G. Zimmer, Jr.	Univ. Tulsa

Discussion, Institute of Metals Division

Contents

PAGE

The Cobalt-chromium Binary System (paper by A. R. Elsea, A. B. Westerman, and G. K. Manning, *Met. Tech.* June, 1948, TP 2393 298

Mechanism of Precipitation in a Permanent Magnet Alloy (paper by A. H. Geisler and J. B. Newkirk, *Met. Tech.* Aug. 1948, TP 2444) 300

Some Effects of Applied Stresses on Precipitation Phenomena (paper by W. L. Finlay and W. R. Hibbard, Jr., *Met. Tech.* Sept. 1948, TP 2470) 301

Diffusion of Carbon in Austenite with a Discontinuity in Composition (paper by L. S. Darken, *Met. Tech.* Sept. 1948, TP 2443) 304

An Electron Diffraction Study of Oxide Films Formed on Nickel-chromium Alloys (paper by J. W. Hickman and E. A. Gulbransen, *Met. Tech.* June 1948, TP 2372) 305

An Electron Diffraction Study of Oxide Films Formed on Copper-nickel Alloys at Elevated Temperatures (paper by J. W. Hickman and E. A. Gulbransen, *Met. Tech.* June 1948, TP 2391) 307

Decarburization of Chrome Nickel Alloys by Their Surface Oxides in High Vacua and at Elevated Temperatures (paper by E. A. Gulbransen, W. S. Wysong, and K. Andrew, *Met. Tech.* Sept. 1948, TP 2438) 307

Fractographic Study of Cast Molybdenum (paper by C. A. Zapffe, F. K. Landgraf and C. O. Worden, *Met. Tech.* Aug. 1948, TP 2421) 308

Effect of Composition on Grain Growth in Aluminum-magnesium Solid Solutions (paper by L. J. Demer and P. A. Beck, *Met. Tech.* June 1948, TP 2374) 309

Effect of a Dispersed Phase on Grain Growth in Al-Mn Alloys (paper by P. A. Beck, M. L. Holzworth and P. Sperry, *Met. Tech.* Sept. 1948, TP 2475) 310

Statistical Rate Theory of Metals—Mechanism of Flow and Application to Tensile Properties (paper by J. W. Fredrickson and H. Eyring, *Met. Tech.* Aug. 1948, TP 2423) 315

Nucleation of Slip Bands (paper by J. G. Leschen, R. P. Carreker, and J. H. Hollomon, *Met. Tech.* Sept. 1948, TP 2476) *and* .. 318

Transient Plastic Deformation (paper by R. P. Carreker, J. G. Leschen, and J. D. Lubahn, *Met. Tech.* Sept. 1948, TP 2477) .. 318

Plastic Deformation of Large Grained Copper Specimens (paper by W. R. Hibbard, Jr., *Met. Tech.* Sept. 1948, TP 2469) 323

Plastic Flow in Anisotropic Sheet Steel (paper by L. R. Jackson, K. F. Smith, and W. T. Lankford, *Met. Tech.* Aug. 1948, TP 2440) 323

The Effect of Thermal-mechanical History on the Strain Hardening of Metals (paper by J. E. Dorn, A. Goldberg, and T. E. Tietz, *Met. Tech.* Sept. 1948, TP 2445) 325

The Room and Elevated Temperature Properties of Some Sand Cast Magnesium-base Alloys Containing Zinc (paper by T. E. Leontis, *Met. Tech.* June 1948, TP 2371) 327

The Low Temperature Properties of Tin and Tin-lead Alloys (paper by H. S. Kalish and F. J. Dunkerley, *Met. Tech.* Sept. 1948, TP 2442) 335

Thermal and Electrical Properties of Ductile Titanium (paper by E. S. Greiner and W. C. Ellis, *Met. Tech.* Sept. 1948, TP 2466) 335

Property Changes during Aging (paper by A. H. Geisler, *Met. Tech.* Aug. 1948, TP 2436) 337

Factors Affecting the Tensile Notch Sensitivity of Magnesium Alloy Extrusions (paper by I. Cornet, *Met. Tech.* Aug. 1948, TP 2419) 338

Effect of Grain Size on Tensile Strength, Elongation and Endurance Limit of Deep Drawing Brass (paper by H. L. Walker and W. J. Craig, *Met. Tech.* Sept. 1948, TP 2478) 339

A Copper-base Alloy Containing Iron as a High-strength, High-conductivity Wire Material (paper by W. Hodge, R. I. Jaffee, J. G. Dunleavy, and H. R. Ogden, *Met. Tech.* Aug. 1948, TP 2422) 340

A High Strength-High Conductivity Copper-silver Alloy Wire (paper by W. Hodge, R. I. Jaffee, J. G. Dunleavy, and H. R. Ogden, *Met. Tech.* June 1948, TP 2366) 340

The Effect of Chromium on the M_s Point (paper by J. B. Bassett and E. S. Rowland, *Met. Tech.* Aug. 1948, TP 2417 341

The Kappa Eutectoid Transformation in the Copper-silicon System (paper by W. R. Hibbard, Jr., G. H. Eichelman, Jr., and W. P. Saunders, *Met. Tech.* Sept. 1948, TP 2441) 343

Stabilization of the Austenite-martensite Transformation (paper by W. R. Harris, Jr. and M. Cohen, *Met. Tech.* Sept. 1948, TP 2446) 344

Secondary Hardening of Tempered Martensitic Alloy Steel (paper by W. Crafts and J. L. Lamont, *Met. Tech.* Sept. 1948, TP 2439) 345

This discussion is TP 2533 E.

The Cobalt-chromium Binary System

By A. R. ELSEA, A. B. WESTERMAN, Junior Members, and G. K. MANNING, Member AIME

DISCUSSION

(G. E. Doan and W. L. Fink presiding)

A. H. SULLY*—There are two points relating to this paper which I would like to discuss.

The first relates to the purity of the alloys. I note that, whereas all the alloys examined were analyzed for nitrogen, no estimations of the oxygen content appear to have been made and, in discussing the purity of the materials, although the authors refer to the oxygen content of the electrolytic cobalt (0.034 pct), they appear to have ignored the oxygen present in the electrolytic chromium. It was first shown by Adcock[8] that the oxygen content of electrolytic chromium may be, and usually is, considerable, the content of chromium oxide Cr_2O_3 being commonly of the order of 1–2 pct. Brenner, Burkhead and Jennings[9] have recently confirmed this fact and have made a comprehensive investigation of the variation of the oxygen content with conditions of electrodeposition. Work in progress at the Fulmer Research Institute on the preparation of high purity chromium-rich alloys has illustrated the difficulty of removing the oxygen content; and we have found that chromium-rich alloys can continue to oxidize when melted under moderate vacuum conditions (5–20 microns). For this reason, chromium-cobalt alloys with up to 20 pct cobalt, made from electrolytic chromium containing initially about 1.6 pct Cr_2O_3 may contain, after vacuum melting, up to 2.5 pct Cr_2O_3. The oxygen content may be determined readily by dissolving the alloy in dilute acid and weighing the insoluble residue of Cr_2O_3. The same method may be used for the determination of the oxygen content of electrolytic chromium, if the chromium is first heated for a period of a few hours at a temperature of about 800°C.

While the oxide content is unlikely to have any serious effect on the phase transformations, it must be remembered that the chromium content of the oxide will be withheld from solid solution and this will have a minor effect on the location of the boundaries in the chromium-rich alloys. A statement by the authors of the oxygen content of their alloys would enable an assessment to be made of the probable precision in the location of the $\epsilon/\gamma + \epsilon$ phase boundary.

The second point which I wish to make concerns the identity of the γ phase. During a preliminary survey of the cobalt-chromium system, my colleague, Mr. T. J. Heal, and I[10] found that this phase was isomorphous with the sigma phase, of the iron-chromium. Andrews,[11] in commenting on our result, has since stated that there appears to be a third isomorphous phase in the system Fe-V. The present investigation places the composition of this phase at the CO_2Cr_3 composition, confirming the earlier work of Wever and Haschimoto.[12] In the iron-chromium system the isomorphous sigma phase forms at the ratio Fe-Cr. In both systems, however, solution seems to be confined to the low chromium side of the relevant stoichiometric composition. The reason for the variation of the stoichiometric ratio for the isomorphous phase between the Co-Cr and the Fe-Cr system is an interesting subject for speculation, and further study of this phase in the two systems may throw further light on the difficult subject of atomic bonding in alloys of the transition metals.

N. J. GRANT* and L. W. KATES*—Several points in this paper that are not quite clear or in complete agreement with our work at Massachusetts Institute of Technology include the following:

The structure of an all-alpha specimen (Fig 4) could be interpreted as a two-phase structure rather than a single phase one. Specimens of chromium contents 10.6, 16.8, 23.0, and 28.8 pct when heated at temperatures of from 1800 to 2200°F (980 to 1205°C) for periods of time ranging from 1 to 48 hr followed by water quenching, invariably showed both F. C. C. and H. C. P. lines on X ray diffraction examination. In addition, these same specimens, when examined metallographically, using such etchants as chrome regia and Marble's Reagent, showed a rather poorly defined but unmistakable two-phase structure. We wondered, in those treatments where the initial structure was "all alpha," if this were checked by X ray diffraction patterns. It was not made clear, but we presumed that the alpha structure was retained at room temperature by water quenching. Our work has indicated that there is a metastable martensitic form of beta which is formed on quenching from within the alpha range which accounts for the H. C. P. lines and the microstructure observed in quenched alpha specimens. If that is the case, the so-called "all alpha" structures are not truly equilibrium structures. It would seem that a true equilibrium alpha structure does not exist at room temperature and that a study of the α to β transformation would best be done by transferring the hot all alpha specimen directly to the aging temperature without the intermediate quenching step.

Although no attempt was made at Massachusetts Institute of Technology to establish the boundaries of the gamma phase field, the presence of gamma phase in alloys as low in chromium as 15 pct seems questionable. It is interesting to note that alloys *V*-20, *V*-26, and *V*-31 were cold worked initially to form beta and all these showed a gamma precipitate when aged for 65 hr at from 600 to 850°C. Alloy *V*-14, however, which had a homogenized structure initially, did not. It seems possible, therefore, that the so-called gamma precipitate is, in fact, some residue of the initial cold-worked structure. This is further suggested by the fact that Alloy *V*-36 which also had a homogenized initial structure showed no gamma precipitate at 600°C, although at 650°C and above, presumably, massive gamma, as exemplified in Fig 8, was found.

We took note of the treatments for developing gamma shown in Table 4 and repeated one of the experiments using our 28.8 pct chromium alloy. The specimen was heated at 1000°C for 1 hr and water quenched and metallographic examination showed the usual two-phase structure. The specimen was then held for 50 hr at temperatures of 875, 850, 830, and 890°C with metallographic and X ray examination being made between each different treatment. The structure in every case was the same, namely a single phase H. C. P. structure. There was no evidence of gamma precipitate similar to Fig 7 resulting from any of the treatments. Several attempts were made to use the etchant described in the paper for identifying the gamma phase; all were unsuccessful. The etchant finally adopted was Marble's Reagent which showed the beta grain boundaries quite well.

Regarding etchants 1, 2, and 3, used by the authors, we never, at any time, were successful in using them as described in the paper on our own alloys. We are not surprised that prior alpha structure did not transform to beta in 50 hr in alloys of less than 20 pct chromium. We did note, however, that a 28.8 pct chromium alloy transformed from alpha to beta com-

---

* Fulmer Research Institute, Stoke Poges, Buckinghamshire, England.

[8] References are at the end of discussion.

* Massachusetts Institute of Technology.

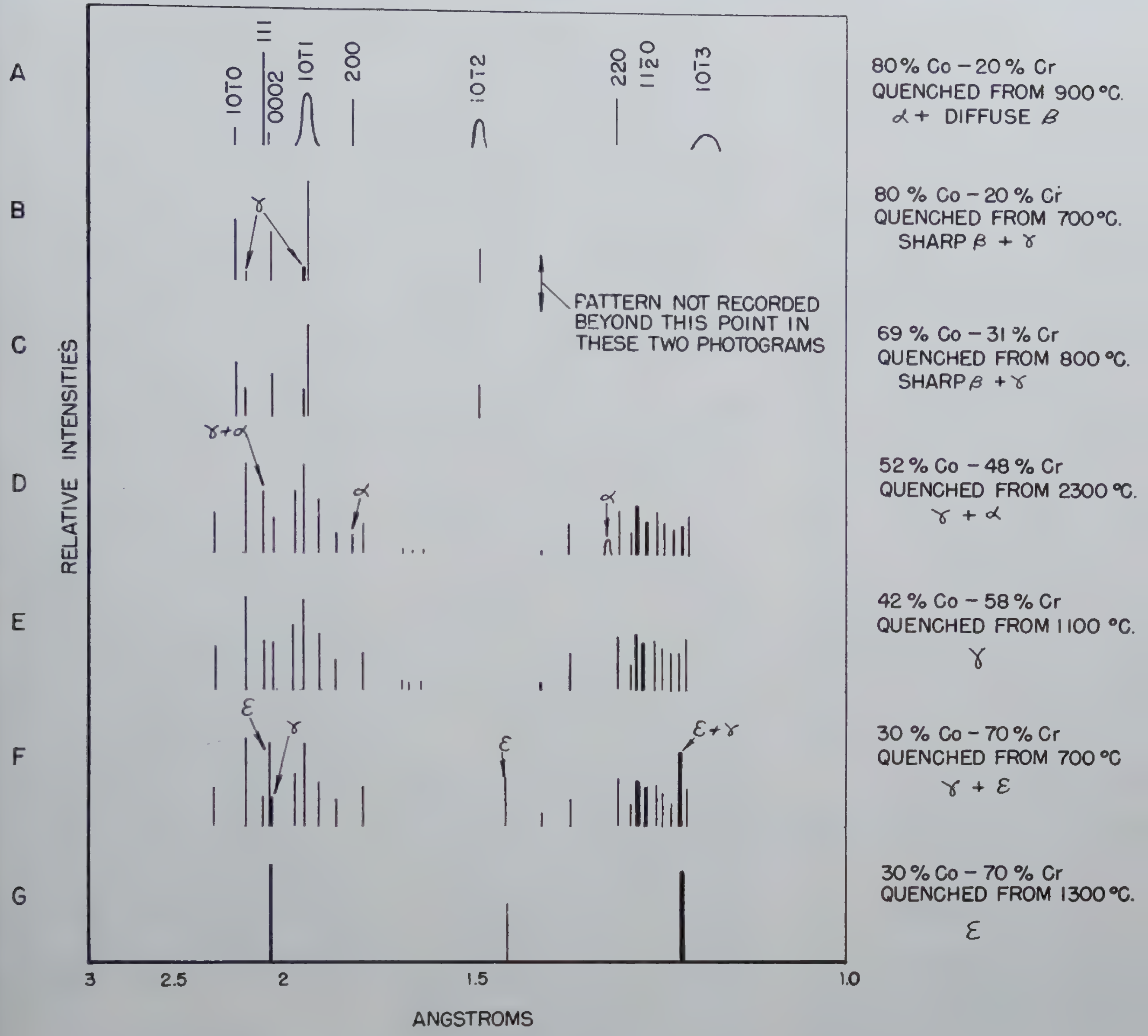


FIG 17—Plots of patterns of various alloy compositions of cobalt and chromium.

paratively easily when cooling in a dilatometer at a rate of about 5°C per min. We found that the 23 and 28.8 pct chromium alloys transformed readily from alpha to beta and vice versa, simply by heating and cooling in a well insulated furnace, but that the 10.6 and 16.8 pct alloys did not transform on cooling (alpha to beta) even at very slow cooling rates.

H. S. AVERY*—Rockwell hardnesses are mentioned in the paper, but I did not see any values. I should like to suggest that the authors will make this more valuable as a reference if in their written reply to the discussions they would include microhardness readings for the various constituents, and also give enough of their X ray crystallographic data to enable other workers to check their data against those described in this paper.

W. O. SWEENY*—The authors should be complimented on their very thorough investigation. We feel it is very timely now because of the necessity to know more about the high temperature alloys, as well as to obtain fundamental data which will enable the development of newer and better alloys for even higher temperatures.

The results of the work which were reported are in substantial agreement with the previous work that has been published and the work done at Union Carbide and Carbon Research Laboratories in Niagara Falls. It is also in line with the work we have done at the Haynes Stellite Co. Of course, the latter has been devoted primarily to more complex alloys.

However, we think it is noteworthy that the results the authors have obtained are not inconsistent with what we found in more complex alloys of this type.

A. R. ELSEA (authors' reply)—Mr. Sully has raised a very interesting question regarding the oxygen content of the cobalt-chromium alloys studied in this investigation. The electrolytic chromium used as melting stock in preparing these alloys contained 0.28 to 0.50 pct oxygen. Alloys *V*-7, *V*-20, *V*-26, and *V*-31, listed in Table 1 of the paper, contained 0.015, 0.14, 0.09, and 0.18 pct Cr_2O_3, respectively, as determined by a wet analysis method similar to the one described by Mr. Sully, but in which a correction was made for the amount of chromium nitride present in the residue. Microstructures of the vacuum-melted alloys, such as those described above, contained numerous oxide inclusions, while the structures of the alloys melted in an alundum crucible under purified argon were almost free from oxide inclusions. After considerable study, it was concluded that chromium oxide in the melt went into solution in the alundum crucible, thus lowering the oxygen content of the melt.

---

* American Brake Shoe Co.

* Haynes Stellite Co.

The observation made by Mr. Sully that the cobalt-chromium gamma phase is isomorphous with the sigma phase of the iron-chromium system substantiates similar observations made at Battelle.

Dr. Grant and Mr. Kates have raised a number of questions regarding phase identification, experimental techniques, and interpretation of the data. In our opinion, the specimen shown in Fig 4 of the paper was alpha phase at the heat-treating temperature used; but upon cooling to room temperature partial transformation to beta phase occurred by a martensite type reaction similar to the reaction observed in pure cobalt by Troiano. This explains the acicular structure observed in Fig 4 and also partially accounts for the diffuse pattern of beta phase observed in the X ray diffraction patterns shown in Fig 17 of this discussion. It was because of this partial transformation during the quench that the aged specimens with alpha prior structures were transferred directly from the preheating furnace to the aging furnace, as described in the paper.

There is no question in our minds regarding the presence of gamma phase in cobalt-chromium alloys containing as little as 20 pct chromium. All of the binary alloys containing 20 pct or more chromium when aged at temperatures 20 to 50°C below the lower limit of the alpha-beta transformation range were found to contain gamma phase. The presence of gamma phase in these specimens was established by both metallographic and by X ray diffraction studies.

Dr. Grant's unsuccessful attempts to use the metallographic etchants described in the paper are difficult to explain. We have used these etchants throughout the study of the binary system and in our studies of the cobalt-chromium-iron, cobalt-chromium-nickel, and cobalt-chromium-nitrogen systems with success. The higher chromium alloys require a much longer etching time, but if the proper etching technique is used, metallographic identification of the phases in this system is not difficult.

The Rockwell hardness impressions referred to by Mr. Avery were made on metallographic specimens to determine the nature of the slip lines or the type of fracture that was produced in the various phases by slight cold work. The microhardness values requested by Mr. Avery are as follows:

Phase	Knoop Hardness (500-g load)
Gamma	940 to 960
Epsilon (84 pct Cr)	630 to 686
Beta (31 pct Cr)	330 to 360
Alpha* (31 pct Cr)	250 to 340

* The hardness of alpha was determined at room temperature and, therefore, it was not pure alpha but was alpha plus beta which formed while cooling to room temperature.

The following discussion of the X ray diffraction data was prepared by Dr. C. M. Schwartz of the Battelle staff, who conducted the X ray diffraction studies:

Plots of representative X ray patterns of alloys of various compositions and heat treatments are shown in Fig 17. Filings were used in Debye cameras with Co, Ni, or CrK radiation; or where greater resolving power was required, block samples were used with CrK radiation in a 20-cm-diam unsymmetrical focussing camera. Since the techniques varied, the photograms could not be compared directly, and it was necessary to resort to the graphical plot in Fig 17.

Pattern *A*, obtained from a 20 pct chromium sample quenched from 900°C, not far above the transformation zone, illustrates the existence of the "diffuse" hexagonal beta phase in samples quenched from the high-temperature F.C.C. alpha region. The corresponding indices of the reflections are indicated above the pattern. The appearance of beta is in agreement with that reported in pure cobalt by Edwards and Lipson.† Some lines are sharp, others diffuse, owing to stacking faults as explained by these authors. The existence of this faulted hexagonal phase is well correlated with the Widmanstätten-like appearance of the microstructure of alpha phase in Fig 4 of the paper. The question whether this structure results in part from quenching strains has not yet been settled from X ray data. Small amounts of "diffuse" hexagonal remain in samples of filings examined in a high-temperature camera at temperatures just above the transformation zone, after moderate holding times. The amount does not increase on rapid cooling to room temperature; however, the strain introduced in this way is not comparable to that in a quenched block sample. Experiments are contemplated in which samples will be held and examined at temperatures far above the transition to see whether the hexagonal structure will tend to disappear.

It is to be noted that the (111) cubic and (0002) hexagonal reflections are resolved; complete resolution is attained in the focussing camera, using chromium radiation. This makes it possible to compare the relative intensities of the lines of the hexagonal phase. In spite of the diffuse nature of some of the lines, there is no marked discrepancy among the intensities, as has sometimes been claimed for cobalt itself.

Patterns *B* and *C* were obtained from 20 pct and 31 pct Cr block samples hot forged in the beta region, and subsequently heat treated and quenched from 700 and 800°C, respectively. Under these conditions sharp hexagonal beta-phase patterns are obtained free from alpha. The two strongest lines of gamma phase also appear faintly, in agreement with the phase diagram in Fig 3. Note again that the relative intensities of the first three lines of beta agree in Patterns *A*, *B*, and *C*.

Patterns *D*, *E*, and *F* contain gamma phase. Pattern *E* is single-phase gamma, with lattice spacings corresponding almost to the limit of chromium solubility. The solubility range is small, and the shift in spacing is readily detected only in the back-reflection lines. Pattern *G*, that of the chromium-rich solid-solution phase epsilon, is given so that its presence in Pattern *F* may be easily identified.

---

† *Proc.* Roy. Soc., (1942) **180A,** 268.

References

8. F. Adcock: *Jnl.* Iron and Steel Inst., (1927) **115,** 369.
9. A. Brenner, P. Burkhead and C. Jennings: *Jnl.* Res. Nat'l. Bur. Standards (1948) **40,** 131.
10. A. H. Sully and T. J. Heal: *Research,* (1948) **1,** 288.
11. K. W. Andrews: *Research,* (1948) **1,** 478.
12. F. Wever and V. Haschimoto: Mett. K. W. Inst. Eisenforschung, (1929) **11,** 293–330.

Mechanism of Precipitation in a Permanent Magnet Alloy

By A. H. GEISLER, Junior Member AIME, **and J. B. NEWKIRK**

DISCUSSION

(G. E. Doan and W. L. Fink presiding)

W. L. FINK*—For some years many of us have believed that the age hardening mechanism consisted simply of precipitation by nucleation and growth in which the precipitation occurred by the usual Widmanstätten mechanism, and during the early stages of which the precipitate is coherent with the matrix.

Dr. Geisler has been working on this problem—one system after another—and he has succeeded in proving that mechanism for several alloys. He is doing a very excellent job. I feel that today he skipped rather lightly over one of the important parts of this work—the streaks that are obtained on the Laue patterns from single crystals. I wonder if he would like to say a little more about the streaks.

---

* Aluminum Co. of America.

A. H. GEISLER (authors' reply)—The interpretation of streaks in Laue photograms represents a specialized but important part of the precipitation problem. It has been demonstrated elsewhere[8] that the streaks can be interpreted generally as the result of small dimensions of the precipitate particles that have a new structure related to the matrix by coherency. As such, the small particles should warrant no special consideration as a step in the precipitation process discrete from the larger, coherent particles that produce normal sharp diffraction. This was suggested by the discusser 8 or 10 years ago and has since been confirmed for several alloys. Indeed, the present study illustrates that age-hardening need not be contingent upon the structure that is responsible for streaks since the major part of the hardening of Cunico occurs after the particles have grown to a size large enough to produce sharp diffraction.

R. M. BRICK*—This aging is a little different from the normal precipitation process. Constitutionally it is almost a eutectoid reaction. You use the word "recrystallization" in discussing the lamellar stage of precipitation. I assume that you imply that if reheated after complete softening, when again in the single phase structure, there would be a completely different matrix grain structure than there was originally. Is that true?

A. H. GEISLER—The difference between the precipitation process in this type alloy and the usual is that here the precipitate has the same structure as the matrix whereas usually the precipitate has a structure different from the matrix. In this respect the reaction here is more remote from the eutectoid reaction where both decomposition products usually have structures different from each other and different from the parent phase. Furthermore, the basic constitutional difference exists: that of the absence of a specific reaction temperature and composition as in the eutectoid reaction. Grain refinement in some alloys can be induced by the nodular reaction. The best evidence for this has been obtained on single crystals using the Laue X ray diffraction technique. Single crystals of some alloys, notably Cu-Be, Ag-Cu and Ni-Au are replaced by a polycrystalline aggregate when heat-treated crystals are aged and re-solution heat treated.

A. E. FLANIGAN*—I should like to comment on Dr. Geisler's remarks concerning a novel method for obtaining grain refinement.

It is implied that in the case of certain precipitation-hardening alloys it has been found possible to obtain grain refinement of the solid solution by means of a process involving precipitation followed by resolution. Presumably, precipitation of the discontinuous "lamellar" type in the region of the grain boundaries of the solid solution is attended by recrystallization of the matrix. Upon re-solution, it is implied, the new orientations of the previously lamellar patches are preserved thus effecting grain refinement of the solid solution.

This mechanism and its possible practical importance interested me several years ago and at that time an attempt was made to induce grain refinement in this manner. A magnesium base casting alloy containing 9 pct aluminum and 2 pct zinc was employed. (ASTM B80-45T grade AZ92.) About ten cycles of the above-described treatment were employed and measurements of the grain size of the solid solution were made after each cycle. No changes in grain size were detected, however, even after this large number of cycles.

A. H. GEISLER—The application of the nodular reaction as a grain refinement method depends both upon alloy and thermal treatment. Since this is a recrystallization of the matrix it can be suppressed by the relieval of strain through prior recovery in much the same manner as the suppression of recrystallization of cold worked metal. Thus, by slow heating to the solution temperature, it has been possible to recover single crystals by reabsorbing the recrystallized nodules. For grain refinement, such a reversion would have to be prevented; the nodular reaction would have to be carried to a sufficient degree of conclusion by aging and the heating rate to the solution temperature would have to be sufficiently fast to avoid reversion to the original grain structure. Naturally, if the rate of grain growth to a stable structure is very rapid at the solution heat treating temperature, as it may well be in some alloys, this method would be of little avail as a possible method of grain refinement.

References

8. A. H. Geisler and J. K. Hill: Analyses and Interpretations of X ray Diffraction Effects in Patterns of Aged Alloys. *Acta Crystallographica*, Nov. 1948, **1**, (5), 238–252.

---

[8] References are at end of discussion.

* University of Pennsylvania.

* University of California.

Some Effects of Applied Stresses on Precipitation Phenomena

By WALTER L. FINLAY and WALTER R. HIBBARD, JR., Members AIME

DISCUSSION

(G. E. Doan and W. L. Fink, presiding)

V. G. PARANJPE*—Have the authors investigated, or theoretically looked into, the effect of these large hydrostatic pressures on the equilibrium diagram itself? Of course, the change in the equilibrium relationships would depend on the amount of heat evolved, and the change of volume associated with the reaction. The effect may be computed by straightforward thermodynamics. It may be that the effect of hydrostatic pressures which the authors describe in their paper might arise appreciably from the change in the equilibrium relationships only.

W. L. FINLAY (authors' reply)—Mr. Paranjpe has raised a very interesting point. We have no information on the effect of pressure on this system and have had no opportunity to investigate it.

I might point out that intermittently through the aging sequence specimens were removed for periods up to two hours in order to make the hardness measurements. Occasionally specimens were examined microscopically and nothing unusual compared to atmosphere-pressure-aged specimens was noticeable under the microscope.

Also, the hydrostatic-aged specimens had hardness properties, which, although they were significantly different from atmosphere-pressure-aged specimens, nevertheless were not of a different order such as no hardening at all. This would make me wonder if the solubility relationship had been markedly changed, but I have no information on that.

C. S. BARRETT*—If you had examined the microstructures, would you say whether there is a tendency for orientation of the precipitates as a result of the applied stresses?

---

* Massachusetts Institute of Technology.

* University of Chicago.

W. L. FINLAY—I am afraid that again we have no data to offer on that point. Optical microscopy gave no pertinent evidence and no X ray work was done in the investigation.

C. S. BARRETT—Did you happen to notice whether the precipitate tended to be aligned on slip planes, or was it distributed generally, and with equal numbers of precipitated particles on the different crystallographic planes that are equivalent in the matrix?

W. L. FINLAY—At the aging temperature involved (100°C), Widmanstatten figures are not developed in these alloys and precipitation particles were aligned on slip planes.

R. M. BRICK*—Do you have any data regarding the volume changes involved in the systems?

W. L. FINLAY—In so far as 100°C aging is concerned, there is a volume contraction in Al-Zu and Al-Cu; data are not available for 75S.

R. M. BRICK—If there is volume contraction, then you would expect pressure to accelerate precipitation just from a volume standpoint?

W. L. FINLAY—Yes, you would expect more precipitate to come out from that standpoint. You are placing the matrix under compression, and, in the Al-Zn case, the zinc is in tension and you are relieving the elastic strain. Therefore, you would tend to extend the amount of coherency. Both Drs. Barrett and Brick have touched on the interesting possibility that the application of external stresses during aging might affect the mode of precipitation, that is, by changing the relative amounts of precipitation on different matrix planes which, although crystallographically equivalent, are differently oriented to an applied uniaxial tensile stress; or by changing the plane of precipitation by altering the relative spacings of matrix and precipitate. An investigation of these possibilities using single crystals would appear to be worthwhile.

G. CHAUDRON,† J. HERENGUEL‡ and P. LACOMBE§ (*Translation*)—The paper by Messrs. Finlay and Hibbard draws attention to the importance of stresses on the kinetics of structural hardening, particularly at a temperature of 100°C where abundant precipitation of a second phase can be observed in most solid solutions of aluminum.

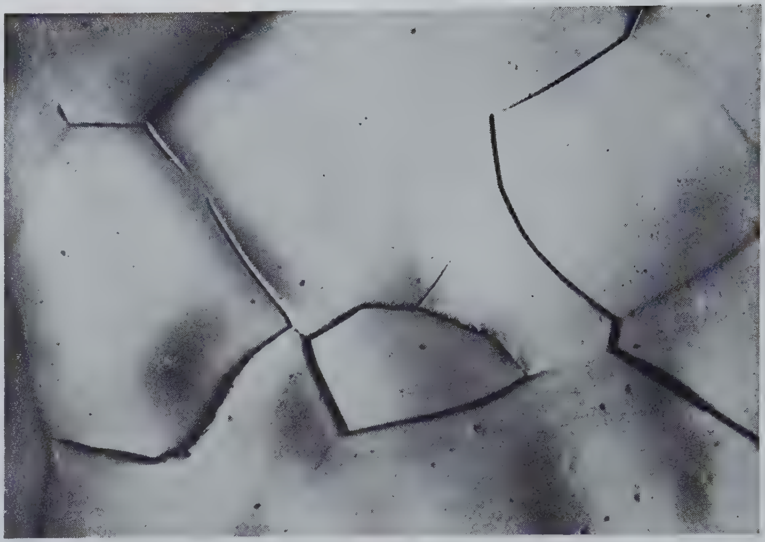


FIG 17—Intergranular decohesion observed spontaneously on aluminum alloy containing 12 pct zinc, electrolytically polished then aged at 20°C after quenching from 400°C. × 300.

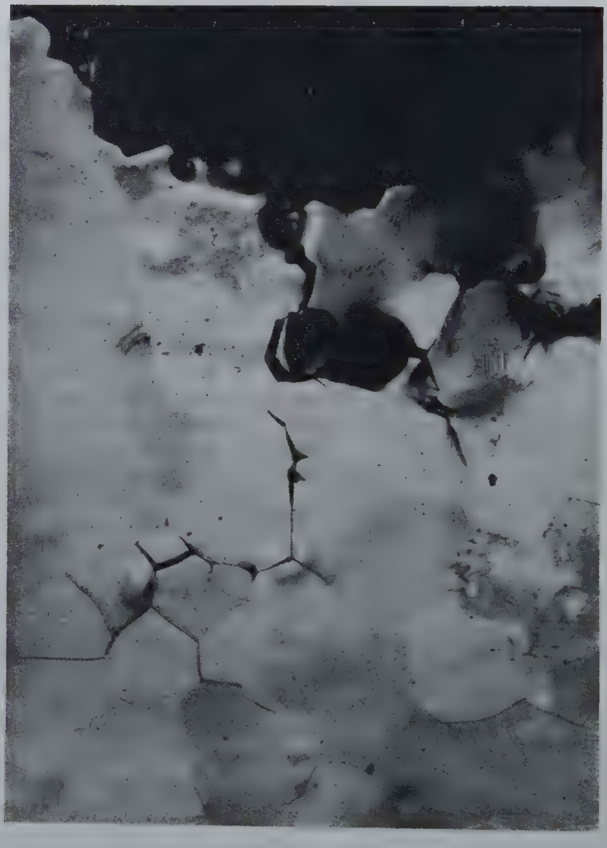


FIG 18—Aluminum-zinc-magnesium alloy hardened at room temperature and then stressed after electrolytic polishing. × 150.

We should like to call attention to some observations which we made with regard to these alloys. We have shown[12] that structural hardening is very rapid at room temperature for an alloy containing 12 pct zinc. Moreover, in the course of a more or less prolonged aging this group of alloys starting with a zinc content greater than 3 pct presents, depending on composition, a very important phenomenon which we have called "intercrystalline decohesion."[13] This phenomenon is characterized by intergranular separation, producing a very great brittleness of the alloy (Fig 17). This brittleness can be observed both in hardening under vacuum and in the presence of more or less humid air. In addition, an external mechanical stress applied to the metal accelerates this cohesion[14] whether the tested specimens are exposed to air or are protected by plating, varnish or anodic oxidation. The resulting fracture is definitely inter-

---

* University of Pennsylvania.

† Director, Laboratory of Vitry-sur Seine (France). National Center of Scientific Research.

‡ Director, Wire Drawing Plants and Rolling Mills Research Center, LeHavre, France.

§ Research Director of Vitry Laboratory, France.

[12] References are at the end of the discussion.

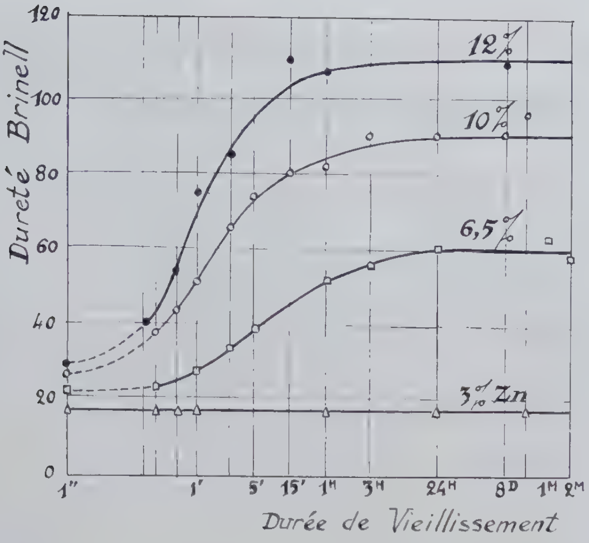


FIG 19—Durete Brinell = Brinell hardness. Durée de Vieillissement = Aging time.

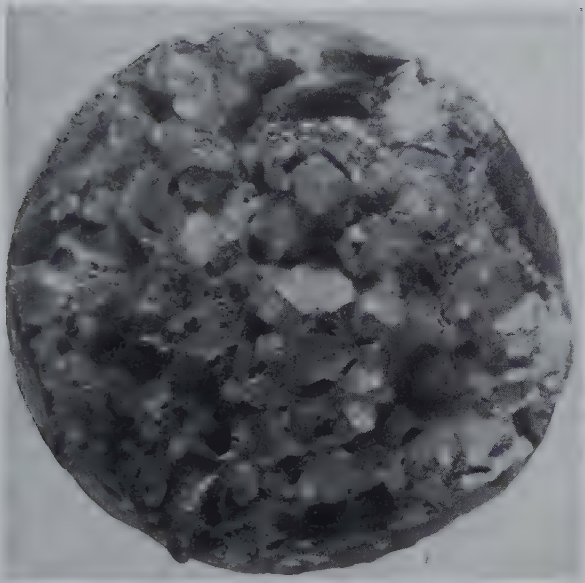


FIG 20—Macrograph of Intercrystalline Fracture of 12 pct Al-Zn after 5 hr at 100°C under 20,500 psi Tensile Stress. Tensile specimen was ½ in. diam. Original magnification 13 ×. Reduced approximately one-half in reproduction.

granular (Fig 18) as found by Finlay and Hibbard at a temperature of 100°C.

As regards the hardness tests made by the authors, we are happy to notice complete agreement with our own tests carried out in 1943 on the same alloys, prepared likewise by starting with pure metals. (Fig 19). With regard to an alloy richer in zinc, 20 pct for example, it becomes impossible to perceive an increase in hardness after quenching, the latter taking place in a few seconds. In the course of the same study, we determined the specific effect of certain impurities, particularly iron and magnesium, on the kinetics of hardening. Additions of iron restrain the intensity of hardening without slowing it up. Magnesium, on the contrary, even in extremely small quantities (0.02 pct) reduces the hardening speed. Thus one realizes that preceding studies made on impure alloys were not able to reveal this structural hardening.

Finally, it seems that after tests made on Al alloys containing 12 pct Zn, Messrs. Finlay and Hibbard intended to associate the phenomenon of retrogression with the particular way of aging under stress. Although the authors confirmed that 200°C was the optimum retrogression temperature for Al-Cu alloys, they adopted for their creep aging tests on this alloy the temperature of 100°C. It seems to us that the choice of this temperature no longer allows comparison of these two alloys from the point of view of the mechanism of redissolving the mass of atoms in the course of retrogression.

W. L. FINLAY (authors' reply)—The interesting discussion contributed by Messrs. Chaudron, Herenguel, and Lacombe is very welcome. The authors were familiar with the publications of these investigators and, in common with a great many others, accord them the greatest respect. The following remarks are offered regarding the various points raised by these discussers:

1. The spontaneous "intercrystalline

decohesion" discussed by Messrs. Chaudron, Herenguel, and Lacombe was never observed by us in an experience comprising perhaps a dozen ingots with compositions from 8 to 25 pct zinc from which several hundred specimens were heat treated and observed on occasion over a two-year period. The discussers' micrographs establish very clearly, however, that their specimens did suffer intergranular fracture, and we have no explanation for the apparent discrepancy. The brittle, intercrystalline fracture we reported is illustrated in the accompanying macrophotograph (Fig 20). Unlike the spontaneous failures reported by the discussers, all of ours occurred after a period of stress at an elevated temperature.

2. The very rapid age hardening at room temperature of quenched high purity aluminum-zinc alloys, shown in our paper and in the discussers' Fig 19, was first reported by Fink and Van Horn* based on both hardness and electrical resistance. Since this feature was only of incidental interest to our main purpose, space did not permit mentioning the gratifying agreement with the work of the French investigators.

---

* W. L. Fink and K. R. VanHorn: Equilibrium Relations in Aluminum-zinc Alloys of High Purity. *Trans.* AIME, (1932) **99,** 132–140

References

12. J. Herenguel and G. Chaudron: *Comptes Rendus* Acad. Sci. Paris, (1943) **216,** 687. *Rev. de Metállurgie* (1943) **41,** 33.
13. J. Herenguel and P. Lacombe: *Métaux et Corrosion* (1944) **19,** 39.
G. Chaudron, J. Herenguel and P. Lacombe: *Comptes Rendus* Acad. Sci. Paris, (1944) **218,** 404.
14. J. Herenguel: *Rev. de Métallurgie* (March–April 1947) **44,** 77.

Diffusion of Carbon in Austenite with a Discontinuity in Composition

By L. S. DARKEN, Member AIME

DISCUSSION

(R. Smoluchowski and M. L. Fuller, presiding)

R. SMOLUCHOWSKI*—Thank you for the very interesting paper. It seems to me that these experiments provide a classical example of the role of activity in diffusion. You mentioned that in your experiment you had a rather wide two-phase region. This is possible in a ternary system only if there is migration of the third component and I do not think that under the conditions of your experiment this was possible.

L. S. DARKEN (author's reply)—The two phase region in the first experiment (Fig 1) does indeed arise by virtue of the migration of the third component (carbon). In a binary system there can be no gradient of activity in a two phase region and hence such a region cannot be obtained by isothermal diffusion. In a ternary system, however, a two phase region is not incompatible with an activity gradient. In fact the data of R. P. Smith demonstrate that in the two phase region (austenite + ferrite), at constant ratio of iron to silicon the activity of carbon rises, though slightly, with inrceasing carbon content.

V. G. PARANJPE†—I would like to ask Dr. Darken if this technique would not be a godsend in estimating the influence of alloying elements on the activity of carbon. For example, the steel containing various alloying elements might be welded on to an alloy in which the activity of carbon is known. The diffusion experiment would thus give an approximate idea about the change in the activity of carbon due to the presence of alloying elements in the former steel.

L. S. DARKEN—Mr. Paranjpe's suggestion seems good in principle. As to whether this method would be easier than others would depend upon the particular circumstances. This diffusion method is inherently longer than the gas equilibration method used by Smith and requires a multiplicity of analyses for each sample. However, the diffusion method does not require such precise control of the gas atmosphere and might be adaptable to a relatively uncontrolled atmosphere—for example, by plating the "sandwich" with an element relatively impermeable to carbon.

J. T. MILEK*—I wonder if the diffusion could also be explained as an ionic mechanism—that is, a movement of the carbon ions, if we could say they exist. I do not know if there is such a thing as a carbon ion in steel.

It was brought out recently that carbon can migrate under the influence of an electric current—a dc current electrolysis. I wonder if you would see this reaction by performing an electrolysis test, instead of waiting 14 days, that is, to impress a high d-c current through the weld and also note that diffusion or transportation of carbon, whether it is ionic or atomic.

L. S. DARKEN—I agree with Mr. Milek's implication that the carbon in austenite is ionic in nature. It migrates under the influence of a potential gradient (direct current); I have observed this effect myself. However, I do not believe anything can be gained in the present type of experiment by the passage of an electric current. In fact, I would expect that the potential gradient would give rise to approximately equal force on the carbon atoms no matter which side of the weld, and that the resulting motion of the carbon would tend to mask the effect considered in this paper. The superimposed effect of the electric current would certainly complicate the interpretation at any rate.

Since the publication of this paper, I have been informed of the prior work of Seith and Bartschat[12] on this subject. They welded an iron-carbon alloy (0.965 pct C) to each of several low carbon alloys (Ni-Fe, Co-Fe, Cr-Fe, Cu-Fe) and held each welded specimen at 940°C for one day. Although "uphill" diffusion was not encountered under these circumstances, a discontinuity in carbon content at the weld was found in several cases. This discontinuity was most pronounced at the weld involving a 19.2 pct nickel alloy—the carbon being 0.39 pct on the high nickel side and 0.66 pct on the iron-carbon side of the weld. A slight discontinuity (about 0.07 pct C) was found at the weld involving a 10 pct cobalt alloy. No discontinuity was found with the copper (2.7 pct Cu) or chromium (10.2 pct Cr) alloy.

---

* Carnegie Institute of Technology.
† Massachusetts Institute of Technology.

* Babcock and Wilcox Co.

Reference

12. W. Seith and F. Bartschat: *Ztsch. für Metallkunde* (1942) **34,** 125.

An Electron Diffraction Study of Oxide Films Formed on Nickel-chromium Alloys

By J. W. HICKMAN, Member AIME, and E. A. GULBRANSEN

Table 6 . . . Indicating Special Significance of Alloying Elements

Group	Useful Life at		Alloy Analysis					Oxide Films at 950°C		
	°C	Hr	−46	Cr	Si	Mn	Others	Fig	Electron Diffraction Pattern	Lines
I	1125	245	133	16.6	1.2		Yes	4	$NiO.Cr_2O_3$	Sharp
	1175	157	120	20	1.3		Yes	1	Cr_2O_3	S
II	1175	86	122	20			Yes	2	$NiO.Cr_2O_3$, Cr_2O_3	Medium
	1125	66	126	16.5			Yes	5	Cr_2O_3	S
III	1175	25	132	20		1.7	No	3	$NiO.Cr_2O_3$	S-Oriented
	1125	21	134	16.2		1.7	No	6	$NiO.Cr_2O_3$	S

DISCUSSION

(R. Smoluchowski and M. L. Fuller, presiding)

F. R. MORRAL*—This paper presents some very interesting data on two types of Ni-Cr alloys. The authors give, it is believed, too much importance to the oxide films formed as being the controllers of the useful life of the alloys at high temperatures. The data, summarized herewith in Table 6 by types of life (more than 100 hr, 50 hr) seem to indicate that alloying elements, such as Si and Mn, are of special significance. The effect of the small amounts of Zr, Ca and Al cannot be properly evaluated. The study of another set of alloys with these elements present and 1.7 pct Mn would be necessary.

The presence of the 1.7 pct Mn is definitely detrimental to the useful life, although the oxide film is the same as that on the alloy showing the longest life (245 hr). Heating and cooling cycles may have an effect on the oxide films, although Fig 7 and 8 show that the high temperature oxides are retained on cooling down to room temperature for all six alloys.

The somewhat larger amount of Cr (3.5 pct) is beneficial even at higher temperatures. This effect was not so pronounced on the alloy containing 1.3 pct Si.

It may be of interest to mention that a salt bath has been found which at 1850°F (1010°C) in less than 10 min. removed the difficulty of eliminating film present on nichrome rods and wire.

The authors are to be thanked for making available this type of valuable data.

E. A. GULBRANSEN (authors' reply) We are indebted to Dr. Morral for his interesting comments. The original choice of heater alloys was made on a historical basis. It would be of scientific interest now to study the effect of one alloying element at a time on both the lifetime and structural factors of the oxide film. This would require a long and intensive study.

We have not made in our conclusions comparisons between alloys of the 80 Ni-20 Cr and the 61 Ni-16 Cr-23 Fe series because the useful life data are made at different temperatures. We agree that the individual effects of Mn and the small amounts of Zr, Ca and Al cannot be separated from the data in the alloys studied. The effect of Si on the useful life and on the structure of the oxide film is very marked in the case of the 80 Ni-20 Cr alloy. The beneficial effect of Cr is also noted.

We would like to restate our interpretation of the oxidation resistance of heater alloys. We feel that the oxide film cannot be considered as a separate entity but rather in combination with the interface and the alloy itself. The processes by which oxidation occurs involve the transport of metal ions from the alloy through the interface and oxide to the surface. The rates of these processes are determined by the structure factors in the alloy, interface and oxide. We also feel that the stability of the oxide to decomposition and solid phase reactions on heating and cooling are of great importance and are not shown by a single cooling experiment.

F. E. BASH*—The work reported by the authors of this paper is very timely in view of the great demand for information leading to the development of better alloys for high temperature service. It is hoped that this work and other work now being undertaken will lead us to a better understanding of the mechanism of oxidation resistance of alloys when subjected to high temperature.

The alloys which were studied in this investigation were prepared in a laboratory high-frequency induction furnace, casting ingots weighing approximately 14 lb. These were forged and a portion rolled to ¼ in. rod and drawn to 0.025 in. wire for life test. The other portion was rolled to strip for this electron diffraction study. Heat 13246 represents the old Nichrome LV (80Ni-20Cr). Heats 12246 and 12046 represent improvements in composition leading to the present Nichrome V (80Ni-20Cr). In the same way, Heats 13446 and 12646 represent the progress of development of Nichrome (60Ni-16Cr-Bal. Fe).

Attention should be called to the life test figures given. It will be noted, for example, that Heat 12046 of Nichrome V (80Ni-20Cr) has a useful life of 157 hr at a temperature of 1175°C (2150°F). This compares with a life of 245 hr at 1125°C (2050°F) for Heat 13346 of Nichrome (60Ni-16Cr-Bal. Fe). It might be thought, in considering these figures, that the Nichrome wire had a useful life approaching that of Nichrome V. It must be understood, however, that the life of a nickel-chromium alloy wire varies as a power of the temperature. For example, 157 hr life for Nichrome V at 1175°C (2150°F) is equivalent to 450 hr at 1125°C (2050°F) for the same wire. We have found, also, that the durability of the nickel-chromium and nickel-chromium-iron alloys is affected by the size of the heat cast. The life on standard production heats of the same alloy composition is considerably higher than on that of the same alloy made in the laboratory furnace. The authors refer to a paper by H. D. Holler,* in which he noted the loss of chromium at high temperatures and the higher temperature coefficient of resistivity of the remaining wire. Due to the increased temperature coefficient, he concluded that the wattage density (and temperature) would be increased in sections of the wire where the chromium was more depleted. I should like to call attention to the effect of varying chromium content on specific resistance and temperature coefficient:

Alloy	Spec. Resis. at 25°C	Spec. Resis. at 1177°C (2150°F)	Av. TC 27–1177°C
80-20 Ni-Cr	650 ohms per cmf	700 ohms per cmf	0.000067
85-15 Ni-Cr	566 ohms per cmf	612 ohms per cmf	0.000072
90-10 Ni-Cr	545 ohms per cmf	585 ohms per cmf	0.000081
A Nickel	60 ohms per cmf	300 ohms per cmf	0.0035

It will be noted from the above that the cold resistance of the nickel-chromium alloys is reduced considerably with reduction of chromium content. It should be noted that the hot resistance also is reduced. The temperature coefficient increases rather slowly with decrease of chromium from 20 to 10 pct. The temperature coefficient of the pure nickel is, of course, very much higher. In view of this information, it is obvious that any areas which may have lost more chromium than the *average* for the wire would not have a greater wattage density (I^2R), since the specific resistance at this point would be lower than the average. Eventually the sectional area reduces more at one point than at others for reasons unknown, and the wattage density increases at this point causing failure of the wire.

---

* American Cyanamid Co.

* Driver-Harris Co.

* Preprint 9207, Oct. 1947, of the Electro-Chem. Soc., "Observations on the Failure of 80 Nickel—20 Chromium Alloy at Excessive Temperatures."

I should also like to call attention to an article by A. H. Sully, entitled "A Simple Method for the Study of Metallic Diffusion in Certain Binary Alloys Systems."[*] Sully determined the concentration gradient of chromium in an 80Ni-20Cr wire 0.021 in. diam after having been at high temperature for a period of time. He combined electrolytic etching techniques with X ray diffraction and thus obtained the chromium content at the surface of the wire and at various depths. He showed the center of the wire to have the original composition, or thereabouts, while the chromium at the surface had decreased to approximately 8 pct. The work of Holler confirms the loss of chromium in prolonged heating of nickel-chromium alloys.

C. L. GUETTEL[†]—I wish to ask a question concerning the nature of the reaction $2Cr + 3NiO \rightleftharpoons Cr_2O_3 + 3Ni$, involving sample 12046, the new Nichrome V. The paper states that this reaction is probable "since Cr_2O_3 is observed both by reflection and transmission above 600°C." However, what becomes of the reaction product, 3Ni? Even if this resulting nickel were immediately oxidized, why would not the NiO be found either in the outer film or next lower oxide layer? Your previous paper states that a spectrochemical analysis failed to detect nickel in the electrolytically removed film.[12] Thus, considering the latter, would not the reaction, $2Cr + \frac{3}{2} O_2 \rightleftharpoons Cr_2O_3$ be more probable?

E. A. GULBRANSEN—I think that both the direct oxidation reaction $2Cr + \frac{3}{2} O_2 \rightleftharpoons Cr_2O_3$ and the solid phase reaction $2Cr + 3NiO \rightleftharpoons Cr_2O_3 + 3Ni$ contribute to the formation of Cr_2O_3 in the film. The Ni ion in the latter reaction diffuses back to the metal. Eventually as the Cr is depleted in the surface of the metal NiO may form at the metal oxide interface. We have confirmed, by X ray diffraction studies of the same alloys, the fact that Ni is not observed in the films at least during the early periods of the reaction.

C. L. GUETTEL—During the discussion period following the presentation of your papers at the AIME meeting of Nov. 1946, I suggested the possibility that a difference in results may occur if the sample were heated electrically instead of by radiation. (My remarks incorrectly were attributed to Dr. C. G. Goetzel). Sully's work on an 0.021 in. 80 pct Ni, 20 pct Cr wire, to which Mr. Bash has already referred, shows that after prolonged heating at high temperature, the remaining metal core nearest to the oxide layer is the most depleted in chromium. The rate and direction of further atom diffusion may be affected in a different manner when an electric current passes through such a core than when the core is heated by radiation. Also, radiation heating places the source of heat on the atmosphere side of the oxide. And electrical heating, in effect, places the heat source on the metal side of the oxide. In view of these comments and the fact that, in actual use, furnace and appliance elements are heated electrically in air, I would like to suggest the following investigation based on the work of Sully and the authors of this paper: Mount and start 16 samples of the same spool of Nichrome V (of known analysis and physical properties) according to the ASTM "Standard Method of Accelerated Life Test for Metallic Materials for Electrical Heating."[13] After one hour, subject two of the samples to the following tests:

1. A temperature determination.
2. A hot resistance determination.
3. A cold resistance determination.
4. Electron diffraction, reflection technique, to determine the identity of the surface oxide.
5. Electron diffraction, transmission technique as applied to the electrolytically removed oxide layer to identify the oxides occurring on the outer surface and in the body of the oxides.
6. X ray diffraction, applied to the remaining metal conductor in conjunction with the electrolytic etching of successive metal layers from the core according to the method of Sully, to determine the transverse chromium distribution at this specific time period.
7. Metallographic examination, transverse and longitudinal sections, to determine the degree and direction of the oxidation.

Repeat the above for other pairs of samples at suitable intervals during the life test. Apply 4, 5, 6, and 7 to the last pair after total life near the point of burnout. It does not appear necessary that 4, 5 or 6 would have to be applied to any sample when heated, as it is unlikely that different oxides would exist at an elevated temperature than when cool, after subjecting each sample to hundreds of two minute heating and cooling cycles prior to the diffraction tests (the one hour samples excepted).

Such a series of tests could relate the formation of the specific oxides with the chromium distribution in the remaining metal core at various time intervals under the standard conditions of the A.S.T.M. life test, and would be supplemented by metallographic information, and thermal and resistance measurements. The same series of tests could be applied to the older types of Nichrome V (80Ni-20Cr) and to experimental modifications of the same alloy. The application of the tests to Nichrome (60Ni-16 Cr-Bal. Fe) would depend on the adaptability of the Sully technique to tertiary alloys.

E. A. GULBRANSEN—Mr. Guettel has outlined a very ambitious program and I think a very good one. We have been doing some thinking along this line but we are not in the position to carry out such a program. We are arranging to make an X ray diffraction study of a number of heater alloys in the following manner: The specimen will be alternately heated and cooled following the ASTM life test procedures. X ray diffraction patterns will be taken of the sample only while the sample is hot. This will be done several times during the life of the specimen. This can be carried out by opening and closing suitable shutters on the X ray tube. A series of patterns also could be taken during the cold cycle of the test. A comparison of the hot and cold pictures as a function of the oxidation may add much useful information to structural changes in the metal and oxide film.

The Sully technique also appears to be a very useful method for studying the change in composition of the metal surface as a result of oxidation.

J. H. SCAFF[*]—I would like to ask the author if it is safe to assume that the composition of the oxide films produced at pressures of 0.1 mm of oxygen would be the same as those oxidized at pressures corresponding to the oxygen pressure of the atmosphere.

E. A. GULBRANSEN—The pressure of 1.0 mm of oxygen is chosen as a matter of convenience in evacuating the camera. In the case of nickel we do not anticipate any change in the structure of the oxide films due to the low pressure. This follows from the fact that NiO is the only oxide ever observed over the temperature range in any work known to us. One point should be carefully noted and that concerns the decomposition of NiO to the metal. If high enough temperatures are reached NiO may not form at 1 mm of O_2 while it would form in higher oxygen pressure atmospheres.

J. T. MILEK[†]—I wish to give my compliments to the authors for their valuable contribution.

I recall some years ago reading Professor Winchell's work on spinels. It seems to me that nickel oxide and chromium oxide form a continuous series of solid solutions in which one would have variable nickel oxide and chromium oxide contents and this should show up in the X ray data. I have not read the paper, and I apologize for that, but I wish the authors would comment on the continuous series of solid solutions and also the replacement of the bivalent end of the spinel.

---

[*] *Jnl. of Scient. Instr.*, p. 244, Dec. 1945.

[†] Driver-Harris Co.

[12] References are at end of discussion.

[*] Bell Telephone Laboratories, Inc.

[†] Babcock and Wilcox Co.

For example, NiO Cr_2O_3—the NiO can be replaced by MnO or MgO, and likewise the trivalent oxide can be replaced by some other oxide as B_2O_3.

Have the authors tried an atmosphere of hydrogen sulphide and what kind of pattern is obtained in a sulphur dioxide atmosphere?

E. A. GULBRANSEN—I failed to mention the problems one encounters in determining the crystal structure of the oxides on the surface. Some of these problems we have mentioned in previous publications. One has the problem of solid solutions of one oxide in another as well as the formation of the spinel structure. The spinel system as Mr. Milek mentioned is a very flexible one in which various oxides can fit. This represents the main difficulty in completely specifying the composition of the oxide. Both X ray and electron diffractions indicate only the lattice spacings and give no direct information about what atoms are in the crystal. The spinels represent a wide variety of oxides which give similar parameters for the common metals used in heater alloys.

To answer your other question, we have not tried atmospheres containing hydrogen sulphide or sulphur dioxide. We have not made such experiments and I would prefer not to comment until we have done so.

References

12. Gulbransen, Phelps and Hickman: *Ind. Eng. Chem.*, Anal. Ed. 18, (1946) 640.
13. A.S.T.M. Test B76-39: Prepared by Committee B-4.

An Electron Diffraction Study of Oxide Films Formed on Copper-nickel Alloys at Elevated Temperatures

By J. W. HICKMAN, Member AIME, and E. A. GULBRANSEN

DISCUSSION

(*R. Smoluchowski and M. L. Fuller presiding*)

J. T. MILEK*—I wonder if Dr. Gulbransen is justified in using the equilibrium constant as a criterion for evaluating which oxide will form. I think, thermodynamically, it is better to use the free energy of formation in his calculations rather than the equilibrium constant and to determine how the free energy of formation varies with the temperature. These thermodynamic relationships might explain in a more logical way why a certain oxide will form in preference to some other oxide than can be explained by the equilibrium constant relationship.

E. A. GULBRANSEN (authors' reply)—The free energy ΔF and the equilibrium constant K are related by the equation

$$\frac{-\Delta F}{2.3\,RT} = \log K$$

It does not make any difference which you use in discussing equilibria.

G. V. PORTER*—At what composition of copper and nickel are there enough electrons contributed by copper to fill the *d* band of the alloy? You have not mentioned in your talk at what combination this occurs.

E. A. GULBRANSEN—The composition is 40 at. pct nickel.

---

* Babcock and Wilcox Co.

* Leeds and Northrup Co.

Decarburization of Chrome Nickel Alloys by Their Surface Oxides in High Vacua and at Elevated Temperatures

By E. A. GULBRANSEN, W. S. WYSONG, and K. ANDREW

DISCUSSION

(*R. Smoluchowski and M. L. Fuller, presiding*)

E. N. SKINNER*—In this paper, Dr. Gulbransen and his associates have employed their usual beautiful experimental techniques to offer a fundamental contribution to our meager knowledge of high temperature gas-metal equilibria. Although the procedures adopted in this work are above reproach, there are several questions which might be raised concerning the interpretation of the data.

It has been considered that all carbon in the three heater-type alloys studied has been entirely in solid solution within 700–910°C, the range of temperatures investigated. This assumption is not consistent with observed micro-structural changes occurring in carbon-containing Ni-Cr and Ni-Cr-Fe alloys. The solid solubility of carbon in alloys of this type is not definitely known but is conceded to be extremely low. For example, carbidic inclusions have been noted after appropriate heat treatment at 450°C in an alloy of 15 Cr, 6 Fe, balance Ni which contained the uncommonly low carbon content of 0.02 pct. A similar base composition containing 0.04 pct will exhibit an unmistakable precipitate of carbide after heating the solution-treated alloy for two hours at 875°C. It is therefore highly probable that the 80/20 and 62/16 samples, containing carbon levels of 0.08, 0.12, and 0.06 respectively, consisted of a carbide-studded matrix at the temperature of testing.

---

* The International Nickel Co., Inc.

If the ribbon upon which the present work was conducted was in the soft condition, this strand-annealing operation probably consisted of rapid cooling from temperatures at or above 1100°C. Subsequent heating to temperatures in the range of 700–910°C after this solution-anneal would be expected to promote the precipitation of carbides preferentially in the grain boundaries. Assuming that this condition existed, the two rates of decarburization noted by the authors might conceivably be explained by a rapid rate of outward migration to the oxide interface by the carbon atoms located in the grain boundaries and along slip and twinning planes and a slower rate of migration through the lattice.

In comparing the decarburization by high vacua vs. hydrogen, we are interested in learning whether the loss in weight of the specimen was appreciably affected by the dissociation of NiO in the Ni-Cr spinel or by loss of residual constituents such as Ca, Zr, Al, Mg, or Si.

E. A. GULBRANSEN and K. ANDREW (authors' reply)—We are in complete agreement with Dr. Skinner about the presence of carbides in the chrome-nickel alloys. This should have been made clearer in our paper. In order to calculate the equilibria of carbon and the oxides in the film it is necessary to assume that the carbon is in solid solution since the thermodynamics of the carbides at the temperatures under discussion is not known. The equilibria calculated with this assumption is an estimate of the actual equilibria which must involve the carbides.

Dr. Skinner has suggested an alternative explanation of the two rates of decarburization noted in the experiments. There is one fact which is in opposition to the mechanism proposed, namely, the second rate is observed after the surface oxide film has been removed. We feel that we are dealing with a reaction between carbon, which is in a large excess in most of the experiments, and two different sources of oxide, the surface oxide reaction with carbon being the more rapid reaction. Such an explanation overcomes the objection noted above.

We have not studied the loss in weight due to dissociation of NiO or loss of residual constituents such as Ca, Zr, Mg, Al or Si.

M. LACHANCE*—Has any conclusion been drawn regarding varied silicon content in an alloy as it affects the decarburized weight losses observed?

E. A. GULBRANSEN and K. ANDREW—We have not studied the decarburization rates as a function of the silicon content in the chrome-nickel series of alloys.

---

* Battelle Memorial Institute.

Fractographic Study of Cast Molybdenum

By C. A. ZAPFFE, Member AIME, F. K. LANDGRAF and C. O. WORDEN

DISCUSSION

(*D. L. Martin and J. E. Burke, presiding*)

H. S. AVERY*—Would you comment on the type of illumination that you use to give the depth of focus in your high magnification photographs?

C. A. ZAPFFE (authors' reply)—In fractography, depth of focus or at least an illusion of depth of focus, is attained by oblique lighting, which serves much as the etch does on polished samples to improve the pattern by varying the reflection from grain to grain. Cleavage facets may be studied visually without oblique illumination, but for photography it is essential.

H. S. AVERY—With your oblique lighting do you use a conventional vertical illumination or dark field illumination?

C. A. ZAPFFE—We have tried both, and by far prefer the conventional vertical illumination. Polarized and other types of illumination, such as "sensitive tint," can also be used. There is no special restriction.

J. H. BOSS*—What kind of microscope do you use?

C. A. ZAPFFE—We used a Bausch and Lomb MILS for this work but have used Zeiss, Leitz, and others. There is no special restriction in type of microscope, except that obliquing has an exaggerated importance.

C. W. HORSTING†—Do you use anything else besides the carbide to deoxidize the molybdenum?

C. A. ZAPFFE—Other deoxidizers could undoubtedly be used, but carbon has the feature of forming a gaseous deoxidation product which can be removed from the system under the conditions of the vacuum.

F. R. MORRAL*—I read this paper with interest. The authors should be congratulated for the interesting and clear fractographs. I note that on p. 3 of their paper only one molybdenum carbide is discussed although there is a possibility that two are present in the MoC diagram as shown by Sykes in the 1948 ASM handbook, p. 1183. I wonder if the MoC would be one of the two discussed in item 4 under the conclusions on p. 21 of the paper.

C. A. ZAPFFE—Professor Morral raises a good point with regard to MoC, which we are glad to accept as he suggests.

---

* American Brake Shoe Co.

* Philadelphia Naval Base.

† Carel, Radio Corporation of America.

* Syracuse University.

Effect of Composition on Grain Growth in Aluminum-magnesium Solid Solutions

By LOUIS J. DEMER, Junior Member, and PAUL A. BECK, Member AIME

DISCUSSION

(D. L. Martin and J. E. Burke, presiding)

C. G. DUNN*—May we have more detail on the time required for recrystallization? How did you determine when recrystallization was complete?

L. S. DEMER (authors' reply)—In order to determine the point of complete recrystallization, the specimens of material were annealed for various periods of time and their hardness values were determined. Then hardness was plotted vs. the time, both to logarithm scales, and it was found that when the recrystallization was not complete, the points fell on a line which had a slope. When recrystallization was complete, specimens annealed for longer periods of time fell on the horizontal line. We took the point of complete recrystallization to be the juncture of these two lines.

C. G. DUNN—The break in the hardness-time curve probably does mean something. On the other hand, it is possible to have recovery of physical properties prior to recrystallization. In other words, the metal may soften without any appreciable recrystallization actually occurring. Furthermore, as grain size increases, there may also be a softening from this source. I am not too familiar with the application of hardness tests to the determination of the course of recrystallization and therefore want the authors' views on the reliability of this method.

L. J. DEMER—This was the method of determining the point of complete recrystallization. Then it was checked by examining the specimens microscopically and those points were taken as points of complete recrystallization.

C. G. DUNN—To make a contrast on the reliability of the hardness test, I would say it does not apply to large grains or single grains of silicon ferrite. For example, we have observed on the basis of a Knoop Hardness test complete softening prior to recrystallization. Therefore, the hardness test may, in my opinion, introduce undesirable complexities into the study of the extent of recrystallization.

P. A. BECK (authors' reply)—We made some X ray diffraction patterns for the aluminum-magnesium alloys, as well as for pure aluminum, and the X ray method confirmed the hardness method as to the time necessary for complete recrystallization.

C. G. DUNN—Concerning another point brought out by the authors I would like more information on the effect of the second phase on the recrystallization process. Your data show a smaller grain size as well as longer time for complete recrystallization with the 2 pct Mg alloy. Unless the induction period is considerably lengthened by the second phase, these two facts would indicate a slower grain boundary movement in the presence of a second phase. As I understand it, the main driving force for recrystallization is internal strain energy. We have here then a grain boundary moving in the presence of a second phase but under the action of strain energy. Could the action of the second phase be discussed further?

P. A. BECK—To my knowledge, no complete analysis has as yet been made of the effect of a dispersed second phase on the two rates pertinent to recrystallization, namely the rate of nucleation and the rate of growth. However, it seems very likely that the main effect of a dispersed second phase in decreasing the rate of recrystallization is exerted through decreasing the rate of growth of the new grains, rather than through an effect on the rate of nucleation.[7]

J. E. BURKE—Would it not develop that all of the observations could be explained by saying that the rate of nucleation was constant but that the rate of nucleus growth had been decreased?

P. A. BECK—Yes, at least qualitatively.

C. G. DUNN—I would like to make one more comment on growth in a strained matrix because the following paper by Professor Beck and coworkers will have more on growth presumably in the absence of strain. If true nucleation occurs, new surfaces are developed and energy is required for this. As long as the matrix is in a strained condition the energy needed for nucleation and growth may be obtained from the matrix. If deformed grains instead of nuclei grow in the strained matrix, the picture is different.

J. E. BURKE—Anderson and Mehl[8] made an observation that is possibly pertinent to this paper. They found that the rate of nucleus growth during recrystallization increased with decreasing grain size of the deformed material. That is, the rate of nucleus growth is greater in an originally fine grained specimen than in a coarse grained specimen, even though the two specimens were deformed by identical amounts. This could indicate that interfacial energy as well as deformational energy, is important in controlling the rate of growth of recrystallization nuclei.

P. A. BECK—I think that there is a more likely explanation of that phenomenon. The fine grained material, when cold worked to the same reduction of area, develops considerably higher hardness than coarse grained material. Thus, in spite of the same geometry, the effective cold work is greater. This accounts for the higher rate of recrystallization.

D. C. JILLSON*—I believe that only one set of fabricating conditions was studied, ending with a final cold reduction of 33 pct. Since changes in fabricating conditions might be expected to affect recrystallization phenomena, and hence might alter some of the conclusions, it might be well to state this limitation in the conclusions.

P. A. BECK—The conclusions are meant to apply under the conditions of the experiments, as fully described under "Experimental Procedure." It appears somewhat impracticable to repeat all those conditions in the "Conclusions."

J. E. BURKE—I have two questions I should like to have the authors comment upon. First, what is the influence of magnesium content upon the average rate of grain growth? The second question concerns the activation energy for recrystallization. Anderson and Mehl reported values ranging from 64,500 cal per mol to 52,100 cal per mol for elongations of from 5 to 15 pct. The present authors report a value of 44,000 cal per mol for a deformation of 33 pct. The range of values encountered serves to emphasize the uncertainty of the physical meaning of the computed "heats of activation." It would seem probable that the change in temperature dependence indicates a change in the atomic mechanism occurring during recrystallization.

The differences in the amount of deformation can account for only a few calories per mole in residual energy left

---

* General Electric Co., Pittsfield, Mass.

[7] References are at the end of the discussion.

* New Jersey Zinc Co.

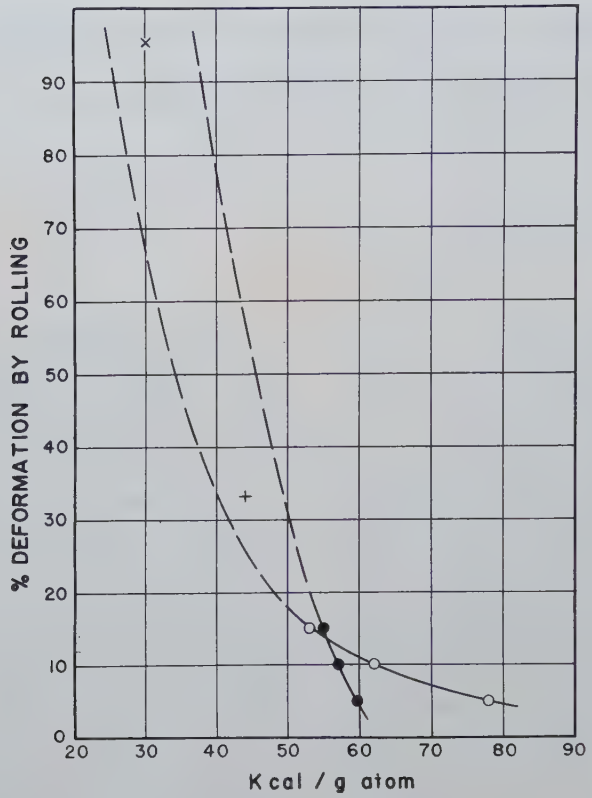


FIG 16—Energy of activation for recrystallization in high purity aluminum.

° Activation energy for nucleation. (Anderson and Mehl)[8]
● Activation energy for growth. (Anderson and Mehl)[8]
+ Overall activation energy for complete recrystallization. (Present paper)
× Overall activation energy for complete recrystallization. (Calvet, Trillat and Paic)[9]

in the metal, and thus cannot account for the difference of several thousand calories per mole in the apparent heat of activation.

P. A. BECK—Concerning Dr. Burke's first question, I should like to refer to Fig 14 and 16 of Ref 1. It is shown there that at small grain sizes the instantaneous rate of growth $\frac{dD}{dt}$ is higher for pure aluminum than for the 2 pct Mg alloy. However, in pure aluminum the rate decreases faster than in the alloy, so that eventually the relative rates are reversed. After about 4300 min. at 400°C, the grain size in the Al + 2 pct Mg alloy was larger, in spite of a smaller grain size, as recrystallized.

Dr. Burke observed the difference between the heat of activation values reported by us on the one hand and by Anderson and Mehl on the other for the recrystallization of pure aluminum. He concluded that this difference emphasizes the uncertainty of the physical meaning of the computed heats of activation. He has, however, neglected to consider the fact clearly shown by Anderson and Mehl, that the *Q* values for both the rate of nucleation and the rate of growth decrease with increasing prior deformation. Fig 16 shows the data from Fig 23 of Anderson and Mehl's paper,[8] and dotted lines indicate the extrapolation of their curves for higher deformation values. Since in our work the two factors, nucleation and growth, were not separated, the heat of activation for the overall process of recrystallization may be expected to lie in between the two extrapolated curves. As seen in Fig 16, this condition is well satisfied, regardless of the uncertainty of extrapolation. The *Q* value estimated from older data of Calvet, Trillat and Paic[9] for 95 pct deformation also falls between the reasonably extrapolated lines. Far from agreeing with Dr. Burke, we feel justified in concluding that, as far as comparisons can be made at all, the data shown in Fig 16 are indeed quite consistent with each other.

References

7. See also Discussion, *Trans.* AIME (1943) **152,** 117.
8. W. A. Anderson and R. F. Mehl: Recrystallization of Aluminum in Terms of the Rate of Nucleation and the Rate of Growth. *Trans.* AIME (1945) **161,** 140, 160.
9. Calvet, Trillat and Paic: Sur la recristallisation de l'aluminium pur. *Comptes Rendus* de l'Acad. Sc. (1935) **201,** 426.

Effect of a Dispersed Phase on Grain Growth in Al-Mn Alloys

By P. A. BECK, Member AIME, M. L. HOLZWORTH and P. SPERRY

DISCUSSION

(*D. L. Martin and J. E. Burke, presiding*)

J. S. SMART, JR.*—From the data presented by the authors, it appears plausible to speculate on the possibility of the existence of a size factor which governs the effectiveness of the precipitated particles in inhibiting grain growth. For instance, the 650° curve of Fig 4 indicates that sufficient precipitate was dissolved in one minute at temperature to remove the inhibiting effect completely. However, the authors noted that the precipitate did not completely dissolve in such short time, but in fact, required something under 125 min. to reach a submicroscopic stage. This behavior seems to indicate that the inhibiting effect is due primarily to particles of very small size which would dissolve rapidly at sufficiently high temperature. If so, it can be readily understood why this curve approximates the behavior of a solid solution alloy despite the presence of the remaining precipitated particles of large size during the whole of the straight section of the curve. Furthermore, this mechanism would explain why the inhibiting effect is strong even though very little visible precipitate is present, as the authors have found elsewhere.

If this reasoning is extended further, the "incubation period" at low annealing temperatures would appear to be the time required for sufficient coalescence to take place to allow the small and effective particles to grow to an innocuous size. At higher annealing temperatures, the incubation period might then depend on a combination of the rates of solution and coalescence of the particles. The apparent anomaly that specimens containing large amounts of manganese compound should have shorter incubation periods, as noted by the authors on p. 14 of the paper, would thus be explainable on the basis that they were in a more advanced stage of coalescence to begin with, and therefore required a shorter period for the coalescence to continue beyond the critical size range.

---

* American Smelting and Refining Co.

The interpretation placed by the authors on the experiments to investigate the effect of various degrees of deformation ascribes a very considerable effect to various degrees of prior reduction. However, from Table 7, it is evident that the process annealing procedure varied from 5 min. at 530°C for the samples reduced 90 pct to 65 min. at 530°C for those rolled 20 pct. This treatment would therefore have resulted in increasing amounts of precipitated compound in the specimens of descending order of reduction. Since the final anneal was at 625°C, which is very close to the solvus point for the alloy used, it does not appear strange to find that the specimens which have the greatest reductions, and the least precipitated compound, lose their restricted status the earliest, as the precipitate dissolves. Is it therefore not unlikely that the differences in process annealing are significant and should be included with differences in reduction in any explanation of the behavior shown in Fig 13 and 14?

P. A. BECK (authors' reply)—Mr. Smart noted the initial slope of the 650°C line for the 1.1 pct Mn alloy in Fig 4 of the paper is approximately equal to the slope for high purity aluminum, as extrapolated to the same temperature from Fig 6 of Ref. 36. From this fact he concluded that in the Mn alloy "sufficient precipitate was dissolved in one minute at temperature to remove the inhibiting effect completely," even though compound particles were found microscopically after much longer periods of annealing. This conclusion is based on the idea that the slope of the log D vs. log t line represents the rate of growth. However, as described on p. 10 of Ref. 36, this is not the case. For any value of D, the instantaneous rate of growth $\frac{dD}{dt}$ depends on the parameter K as well as on the slope n. If a grain growth line is obtained for high purity aluminum at 650°C by extrapolation from Fig 5 of Ref. 36, it is easily seen, by comparison with the 650°C line of Fig 4 of the present paper, that actually the instantaneous rate of growth for high purity aluminum is about 200 times larger than that for the alloy, compared at the same grain size. This suggests that the manganese compound particles seen under the microscope are in fact very effective in reducing the rate of growth. There appears to be at present little basis for the suggestion that the inhibiting effect is due primarily to particles of very small size. Actually, it is expected that any difference in inhibiting effectiveness that may be ultimately traced back to a particle size effect, would turn out to be rather small. Every indication points to the fact that it is the *total number* of dispersed second phase particles that primarily determines the effectiveness of inhibition. It is, of course, obvious that if the *same total amount* of precipitate is present in the form of smaller particles, the effect is larger, simply because the number of particles is then larger.

Mr. Smart is correct in observing that the series of specimens in Table 7, prepared for the study of the effect of increasing prior deformation, did not receive identical processing. The situation was, however, not nearly as unfavorable as it appears from consulting Table 7 only. As shown in Table 3, the treatment previous to the varying 530°C anneals of Table 7 was a series of anneals at 550 and 530°C common to all specimens. As a result, the differences in processing the specimens of series C, D, E, F and G, had no significant effect on their coarsening characteristics. This was also shown in a later experiment, not included in the present paper, where the attempt to finish with the same specimen thickness was given up, and instead the processing treatment was kept constant. In this experiment the decrease of the incubation period with increasing reduction by rolling was confirmed.

C. G. DUNN*—I want to congratulate the authors on their excellent results. They certainly are adding fundamental knowledge to grain growth processes, and this we need. The effect of dispersed phases is particularly important, I think. Using the authors' term "coarsening," it perhaps should be mentioned that phenomena other than the effect of a dispersed phase can be associated with coarsening, particularly in fine grained materials where textures play an important role. Would the authors care to say something on this aspect of coarsening?

P. A. BECK—Dr. Dunn, knowing about our latest results, is offering us here a welcome opportunity for presenting them in this discussion.

Cook and Macquairie[44] showed that in tough pitch copper two different types of coarsening may occur. The one type, which they produced by annealing at 750 to 850°C, after about 50 pct reduction by cross rolling, did not occur in oxygen-free varieties of copper. It was recently found[45] that this type of coarsening, which is characterized by fairly round coarse grains with irregular, "ragged" boundaries and small, completely enclosed twins (Fig 31), may be obtained in tough pitch copper after straight rolling, as well as cross rolling. It appears very likely that this type of coarsening depends on the presence of finely dispersed Cu_2O particles in the copper, and that it is, therefore, closely related to the coarsening phenomena in Al-Mn alloys, described in the paper. The term "random coarsening" used in the introduction of the paper, is descriptive only in the sense that these coarse grains *may* lack any preferred orientation, and that this type of coarsening does not depend on the presence of a texture in the fine grained recrystallized matrix. However, as shown by Bowles and Boas[46] in a paper published after the present paper was submitted for publication, it is possible for the coarse grains growing in commercial aluminum (which contains dispersed phases) to possess a preferred orientation. For this reason it now seems preferable to describe this coarsening as "*inhibition-dependent*" rather than "random."

The other type of coarsening in copper found by Cook and Macquairie, and before them by Dahl and Pawlek,[29] does not require Cu_2O particles. It may occur in oxygen-free copper as well as in tough pitch copper. What it does require, is a well-developed "cube texture" in the matrix, as recrystallized. In the presence of such a sharply defined texture the small recrystallized grains, most of which have fairly closely the same orientation, show great reluctance to undergo gradual grain growth. Copper with the "cube texture" remains quite fine grained up to high temperatures. Finally at about 1000°C (or at a lower temperature after a longer period of annealing) it abruptly develops very large grains. Fig 32 shows such grains embedded in the fine grained matrix in which they grew. Typical of the coarse grains of this type is the presence of very long, straight twin boundaries, inclined at about 45° to the direction of rolling (Fig 33). These coarse grains have a well-developed preferred orientation, different from that of the fine grained matrix. The orientation relationships were recently studied by Bowles and Boas.[46] It was recently found at this laboratory that, even in high purity aluminum, gradual grain growth was practically prevented in the presence of a very sharply defined texture,[47] regardless of the orientation with respect to the rolling direction. Apparently, the determining factor here is the similarity of the orientation of neighboring grains of the fine grained recrystallized matrix. When high purity aluminum with such a structure is annealed long enough at 500–600°C, abrupt coarsening sets in. Grains many times larger than the specimen thickness may be obtained by this mechanism.

The views expressed in the introduction of the present paper are thus confirmed: there are at least two different types of coarsening, the one depending on a dispersed second phase, the other on the presence of a very well developed texture. The common feature of both types of coarsening is the suppression of the continuous or gradual type of grain growth and the abrupt development of extremely large grains which grow by feeding

---

* General Electric Co.

[44] References are at the end of the discussion.

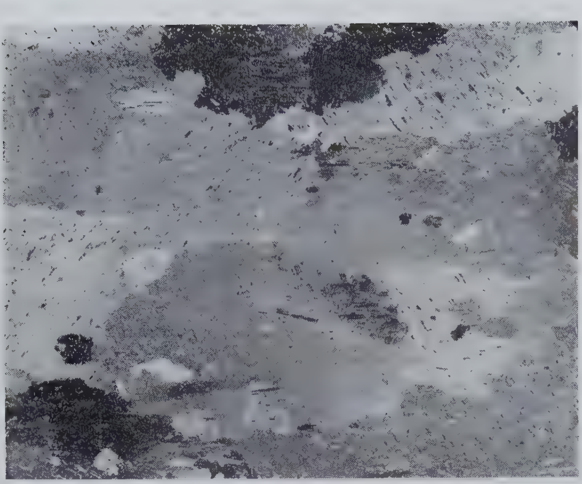


FIG 31—Tough pitch copper after final rolling of 50 pct and final anneal of 1 hr at 800°C.

The coarse grains have rounded shapes and ragged boundaries. Numerous small twins are entirely surrounded by coarse grains. Magnification × 5. Reduced one half in reproduction.

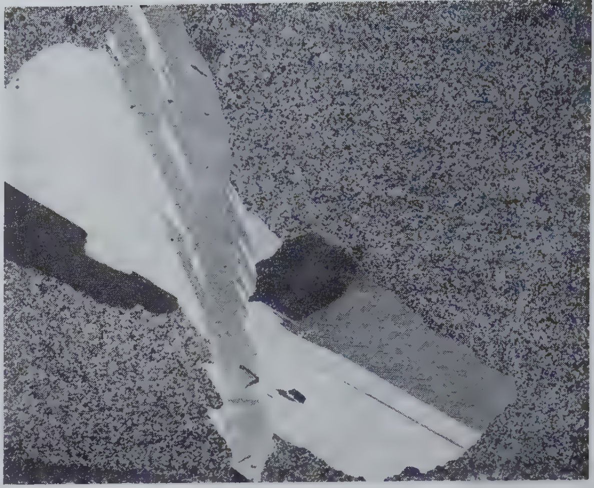


FIG 32—Same copper as in Fig 31, cold rolled 90 pct, annealed 3 min. at 1000°C.

Coarse grains in fine grained matrix. Magnification × 5. Reduced one half in reproduction.

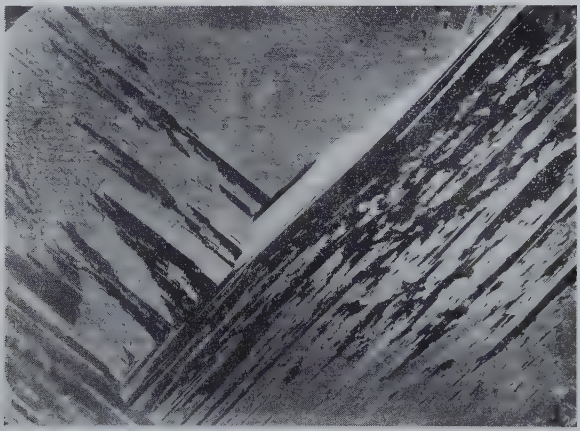


FIG 33—Same copper as in Fig 31, after final rolling of 90 pct and final anneal of 25 min. at 1000°C.

The appearance of this structure is dominated by the characteristically long, straight (twin) boundaries. Magnification × 5. Reduced one half in reproduction.

directly on the fine grained matrix.

C. S. SMITH*—It is, perhaps, somewhat undesirable to use the term "attraction," as do the authors on p. 25 of their paper, for the interaction between the grain boundary and a particle of the second phase. Surely, though a boundary will cling to an inclusion once it has met it, and can be retained in positions that involve considerable curvature, it will not be influenced by an inclusion at a distance.

The effectiveness of such an anchor is relatively little dependent on the surface energy of the inclusion. It is essentially a question of geometry, for the numerical value of the interfacial tension of the boundary cancels out, being both the force for driving and for restraint. If the second phase is not rigid, it will become lenticulate in shape under the influence of the triangle of surface forces around its edge, but such departures from a spherical shape will increase the length of the line of contact, and hence the restraining force, by factors of not more than two or three except when the dihedral angle approximates zero and the second phase spreads as a film separating the grains.

I have come to believe strongly in the simple geometric theory of grain growth, first clearly enunciated by Harker and Parker. The initial contact between grains will produce a random network of surfaces and points of contact. Each internal intersection must involve three grain boundary surfaces meeting along a line and six grain boundaries (and four edges shared with four grains) meeting at each point. If all surfaces exert the same surface tension they will tend to establish angles of 120° with each other and the edges will tend toward the stable "tetrahedral" angle of 109°20′ with each other. Except with ideal stacking of minimum-area tetrakaidecahedra (the three-dimensional equivalent of hexagons), curvatures are inevitably introduced to reconcile the required angles with the random points of contact and this introduces instability. The surfaces will then tend to move toward their centers of curvature, allowing further continual readjustment of the edges. This will proceed smoothly (at a rate dependent on the temperature and the local curvature) until more than three surfaces meet together at an edge or more than four edges meet at a point. The moment this happens the system becomes locally highly unstable and the surfaces will rapidly assume new positions of local equilibrium. Occasionally there will be formed tetrahedral grains, all of whose sides are convex and therefore bound to move inwards, bringing the points of contact with the four neighboring grains together in a stable balance as the grain disappears. Only as this happens will the number of grains decrease or the average grain size increase. This is illustrated two-dimensionally by the accompanying figures. The boundaries of the five-sided and four-sided grains shown in Fig 34A will move progressively toward their centers of curvature through the positions shown by the dotted line, to reach eventually approximately the configuration shown in the solid lines of Fig 34B. If the surfaces are all of equal energy, the moment four of them meet they become highly unstable, for, in place of the original six equilibrium angles of 120°, there are now two angles of 120° and two of 60°. The slightest disturbance will cause the films to seek new equilibrium with a new interface forming at approximately 90° to the one that has disappeared, and there will be accompanying readjustments throughout, as shown by the dotted lines in Fig 34B. The net result is that the five- and four-sided grains have become respectively four- and three-sided, with the addition of one side to each of two neighboring grains. The three-sided grain can now disappear completely by its sides moving toward their centers of curvature without any further instability occurring.

---

* University of Chicago.

If something should happen to render the intersection of four surfaces or five edges stable, further grain growth could not take place. The situation that normally is followed by the most rapid motion now constitutes a complete block. Such an obstacle can be caused either by inclusions (where the requirement is only that all boundaries meet its surface at 90°) or by the grain boundary energy differing from boundary to boundary. In a single-phase alloy, although the energy is clearly not highly sensitive to orientation over most of the range of mismatch, it is likely—in fact, the recent experiments of Dr. C. G. Dunn make it certain—that in the range of slight mismatch there is a high dependence on orientation. Consider the grains shown in Fig 34B with the four interfaces momentarily meeting at a point. This is unstable if the four interfaces and the new interface that must be produced all have equal energies. If, on the other hand, the new interface (the horizontal one in the figure) should have a higher interfacial tension than the resultant of the other four tensions in its direction (that is, if it has a value greater than the diagonal of the polygon of forces they produce) the four films will be quite stable in contact with each other, unless and until movement of other parts of the boundary allows changes of the local angles to such an extent that the above criterion is no longer true. Such four-grain corners are frequently seen in two-phase structures (for example in the case of alpha-beta brass, where the grain boundary between two alpha grains has about 1.3 times the energy of the other possible interfaces) and it can occasionally be found in normal single-phase polycrystalline metals, particularly if the metal possesses a high degree of preferred orientation. The formation of these stable points can completely stop grain growth unless the grain size is extremely small and the curvatures high. Such a blocked network of grain boundaries will be quite stable on indefinite annealing unless some other action sets in to release it, whereupon—if it occurs locally—one grain may become much larger than its neighbors and continue to grow, for a former block will no longer be effective against the new, high, local curvatures. This sudden removal of a block could occur if it were possible for grains to change their orientation under the influence of surface forces (or if orientation changes as a result of local plastic deformation or new nucleus formation), with resultant change of interface energy and upsetting of the local force-polygon. If blocking is due to inclusions, the solution of the finest particles would have a similar effect. In either case, some grains would become considerably larger than their neighbors, the effective negative curvature of their boundaries larger, and thereafter their boundaries would progress continually and at a more or less constant rate into the network of smaller grains in which motion remains blocked. This will continue until the growing grain encounters another one of the same type. The frequency of nucleation will determine whether the grains are only a little larger, as in the case of abnormal grain growth in steels, or enormously larger, as in the general case of "secondary recrystallization."

In addition to blocking at a point by inclusions or by the necessity for the formation of a high-energy interface, it is also possible for a grain boundary to be blocked away from an intersection by inclusions, or by surface forces alone if the boundary, moving toward its equilibrium position, has to change its angle and pass through a minimum energy direction in relation to the two lattices that it divides. Unless the curvature is great enough to overcome this, the boundary will have a definite tendency to "hang" along certain preferred directions between the two crystals. In practice it is probable that the stopping of grain growth is due to a combination of all these factors. It does seem to be a fact that "secondary recrystallization" occurs only in material that has substantially ceased growing in the normal fashion. It is an interesting speculation as to whether primary recrystallization is also a result of a somewhat similar reaction taking place in a much finer network of interfaces.

Although a phenomenon as complex as grain growth clearly cannot be the result of a simple system of causes, it is nevertheless possible to account for many of the observations on the simple basis of the geometry that would result from various interfaces interacting in accordance with the principles enunciated by Plateau, Gibbs, and others in the nineteenth century.

P. A. BECK—Dr. Smith's interesting contribution to the discussion is much appreciated. In the absence of plastic deformation, grain boundary migration in pure metals, resulting in a gradual increase in the average grain size, is undoubtedly motivated by the interfacial tension connected with the grain boundaries. This view was clearly expressed by H. Althertum[48] about 25 years ago. The detailed theory of the geometry of grain boundary migration, as stated by Dr. Smith, was developed more recently by Harker and Parker.[49] Although the latter authors used an atomistic terminology, it has been certainly quite clear that they investigated the behavior of grain boundaries in single phase metals under the influence of what is phenomenologically described as surface tension forces. Their considerations are, in the main, doubtless correct, but require two corrections. First, the relative lattice orientation of neighboring grains affects the value of the interfacial tension.[50] This brings about large deviations of the equilibrium dihedral angles from 120°, depending on the orientation. Also, it is becoming increasingly clear that the simple surface-tension theory of grain growth needs to be completed by consideration of the effects of orientation, probably even beyond those just mentioned. Secondly, although the grain boundary movements serve the tendency to approximate equilibrium angles and straight boundary surfaces, the deviations and irregularities are nevertheless self-perpetuating. Stable grain boundary geometry is thus not actually reached through grain growth, as was originally postulated by Harker and Parker. It was clearly shown by recent isothermal grain growth studies with high purity aluminum[36] that grain growth continues until stopped by the outside surfaces of the specimen.

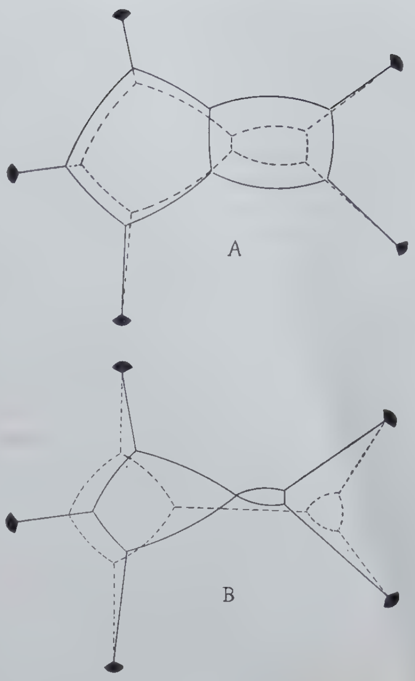


FIG 34—Tracings of soap bubbles illustrating idealized shapes and adjustment of grain boundaries moving under the influence of surface tension.

It is gratifying that Dr. Smith agrees with the view set forth in the present paper, that coarsening is motivated by essentially the same forces that produce gradual grain growth. Whether the coarsening results in the formation of grains only ten to a hundred times larger than the fine grains of the matrix, as in steels, or 10^3 to 10^4 times larger, as it does under certain conditions in copper, the process is always actuated by surface tension, modified by the requirements of orientation. Terms like "abnormal grain growth" and "secondary recrystallization," which were somewhat discredited by ambiguous usage, tend to confuse this picture.

As described in the paper, *inhibition-dependent coarsening* is a type of grain growth that starts at isolated points in a fine grained structure, after general gradual grain growth had been halted by the inhibitive effect of a dispersed second phase. In the examples described, the inhibition was locally overcome by the re-solution and coalescence of particles of the dispersed phase. Once a certain grain, through favorable local conditions, succeeded in growing considerably larger than its fellows, it will continue to grow with relative ease at the expense of neighboring small grains. This effect of "grain size contrast" was found long ago by Z. Jeffries.[1,2]

It should be emphasized here that the above outlined mechanism for coarsening does not require any preferred orientation of the small grains of the matrix. Neither could it conceivably lead to a preferred orientation of the coarse grains formed from a random matrix. However, the mechanism considered may be still operative if the fine grains of the recrystallized matrix did possess a certain amount of preferred orientation. In such a case, it would be expected that all grains which happened to have local conditions of inhibition favorable for growth would not be, in general, equally likely to grow. Those having a particularly favorable orientation relative to their neighbors will be preferred. Thus the presence of a preferred orientation in the fine grained recrystallized matrix leads to a preferred orientation of the coarse grains. This condition is simply a result of the dependence of the rate of growth of a grain upon its orientation with respect to the preferred orientation of its neighbors. There is no need to assume the formation of "nuclei" of a certain orientation, or to postulate that residual portions of the deformed structure remain to serve as nuclei after surviving recrystallization, general grain growth and long periods of recovery at high temperatures.

Although preferred orientation of the coarse grains *may occur* in inhibition-dependent coarsening, it is by no means a necessary condition. On the other hand, in the case of *texture-dependent coarsening*, as explained above in the answer to Dr. Dunn, the fine grained recrystallized matrix *must* have a sharply defined texture in order to restrain gradual grain growth; therefore, the coarse grains growing in this highly oriented matrix have, of necessity, a preferred orientation of their own. The specific mechanism proposed by Dr. Smith for the restraint of gradual grain growth in a fine grained metal with a strong texture, is interesting. Whether or not it actually accounts for the observed facts remains to be established by further research.

J. P. NIELSEN*—Referring to Dr. Smith's discussion on interfacial energies between grains not identically oriented, I wonder if, in the case of close-packed hexagonal metals, a new and perhaps important factor does not also enter into the picture: the difference in thermal expansion coefficients between two adjacent grains at the surface contact. Such a factor would depend of course on the increase in temperature above that at which the metal is completely relaxed.

C. G. DUNN—In the low range of orientation differences, I believe we actually need more data. This work on grain boundary surface tensions, which I hope to report at the next meeting of the Institute, actually was somewhat exploratory in nature. The groups of crystals investigated—each a three-grain group as was described in a technical note[51]—had crystals with a (110) plane in the plane of the sheet. The only variable, therefore, was the direction of the cube edge in the plane of the specimen.

Since the orientations were controlled, it was possible to have two grains in each three-grain group with the same difference in orientation. An arbitrary value of 15 was used. With this common orientation difference providing communication between groups, an entire range of orientation differences could be investigated.

Omitting a description of the entire curve, the shape of the curve indicated values of surface tension at orientation differences of 6 to 8° that were about one half the maximum value. The shape of the curve in this range of orientation differences is important to the understanding of strong texture materials such as copper and nickel-iron with the "cube texture." Strong texture materials should have less grain boundary energy than weak texture materials with the same average grain size.

So much for the surface tension measurements. I would like to say something about orientation relations as a factor in growth processes. First, there is the well known principle that a difference in orientation is required for growth. Secondly, certain orientation relationships, as for example twins, appear to be less favorable than other orientation relationships for growth.

With orientation relationships in mind let us consider the behavior of fine grained copper or nickel-iron with a strong cube texture. What grows in these materials by coarsening or secondary recrystallization always has an orientation which is different from that of the cube texture. Pawlek in 1935 reported that the orientation of the large grains in nickel-iron were as follows: a (120) plane in the plane of the sheet and a [001] direction parallel with the rolling direction. Similar results have been reported for copper but there are also other textures. Omitting twin orientations (actually appreciable material is present as twins in the large grains) there are still many definite preferred orientations remaining. From symmetry they can be reduced in number and classified in relation to the cube texture. All fall 20 to 30° away from the cube texture. One orientation relationship reported recently by Bowles and Boas in Australia has a (111) plane common with the cube texture.

If we call coarsening of the grain structure, starting with a strong texture fine grain matrix, as "strong-texture dependent coarsening" in contrast to "dispersed phase inhibited coarsening," we still need another type designation to describe the case of development of the strong (110) [001] texture in silicon iron.* Unlike copper and nickel-iron, the final large grained silicon iron sheet, to which I have reference, has a strong single texture. Furthermore, the large grains do not have any unusual characteristic shapes or shapes with straight boundaries as Professor Beck pointed out is the case for oriented coarsening in copper and nickel-iron. The prior fine grained material in the as-recrystallized condition does not have a strong single texture as does the copper or nickel-iron examples but has a rather weak texture composed probably of two or more weak preferred crystal orientations. The development of large grains in silicon iron, therefore, is quite different from the development of large grains in materials having the cube texture. Nevertheless, I think it could be classified also as "texture-dependent" but certainly not as "strong-texture dependent."

P. A. BECK—Dr. Dunn's results, and his method for determining interfacial energy as a function of the relative orientation of the adjoining grains are indeed very interesting and pertinent to the questions discussed. We have also noted the fact, referred to by him, that the preferred orientation of the coarse grains usually deviates about 20 to 30° from the preferred orientation of the fine grained matrix. It would be interesting to speculate on the possible causes of preferential growth with an orientation difference of this particular range.

On the basis of information now available to us concerning the coarsening in silicon-iron, it appears that this is not of the "texture-dependent" type. It may be suspected, however, that it depends on inhibition by at least one dispersed second phase. With a complex material of commercial purity, which probably contains several dispersed phases, the exact role of each is difficult to ascertain. But it seems that the dispersed manganese sulphide, which is removed during annealing, may have an important role in the process. It was shown recently[52] that under certain conditions nitrogen (perhaps in the form of dispersed aluminum nitride particles) may provide so much additional inhibition that coarsening is prevented altogether. The role of the much smaller amount of nitrogen normally present in silicon ferrite has not yet been clarified. Possibly, the coarsening reaction depends on the combination of two or more inhibitors, in analogy to the double inhibitor used with tungsten, as described in the Introduction of the present paper. In this connection it should be mentioned that, in the method most generally used for the manufacture of coarse grained tungsten filaments, the inhibitor used is potassium silicate. Most of this inhibitor evaporates during the first heating of the filament, while the coarse grains are formed. However, a portion of it remains in the filament. Here, too, the situation is quite similar to that described in the paper for the coarsening of Al-Mn alloys. As explained above in the answer to Dr. Smith's discussion, preferred orientation of the coarse grains may occur under conditions of inhibition-dependent coarsening, so that the phenomena in silicon-iron do not appear to necessitate a separate class of coarsening.

---

* College of Engineering, New York University

* The (110) [001] texture was discussed at the 1948 ASM Symposium on "Cold Working of Metals" under the subject "Recrystallization Textures."

J. E. BURKE*—The authors have presented an excellent paper and have demonstrated very clearly the importance of growth inhibiting dispersed phase in promoting the appearance of very coarse grains. I should like to present a partial explanation of the role of the inhibiting phase. As Zener has shown, and as the authors demonstrate in this paper, there is a definite relationship between the concentration of the inhibiting constituent, and the limiting grain size, or limiting boundary curvature. Grain growth will cease when the driving force for grain growth, the curvature of the grain boundaries, is just balanced by the inhibiting effect of second phase inclusions.

Now, considering the situation in two dimensions, all grains are composed of sides and corners, and at the corners three boundaries usually meet at angles of about 120°. If all the grains have six sides, the sides also can be flat, and as Harker and Parker[49] showed, there will be no tendency for grain growth. If the boundaries always meet at corners at angles of 120°, the curvature of the faces will increase as the number of sides on a grain increases, being concave outwards for more sides than six and concave inwards for less sides than six. In a reasonably uniform aggregate of grains, the number of sides on grains will vary from three to about eight or nine. In these cases, no very large curvatures are encountered. If now, due to some fluctuation, a much larger grain is produced, it will have many sides, being bounded by small grains. These sides will have more pronounced curvatures than those of more or less uniform grains. Thus, the amount of second phase impurity that inhibits the migration of grain boundaries in the uniform matrix will be insufficient to prevent the migration of the boundary between the larger grain and its smaller neighbors, and the large grain will consume its smaller neighbors. The rate of migration of this boundary is presumably proportional to the curvature of the individual curved sections. Since this is controlled almost entirely by the size of the small grains, the boundary should migrate at a uniform rate. Furthermore, the large grain will continue to grow until the small grains are entirely consumed, and it contacts another large grain. Growth will then stop. The ultimate grain size will thus be controlled solely by the number of large grains that appear.

This argument does not explain the nature of the fluctuation that leads to the appearance of the initial large grains. It could result from the local disappearance of a few inclusions, as the authors suggest, which would permit the local appearance of a very coarse grain. Any other cause for producing a large grain should lead to the same final sequence of events.

P. A. BECK—Dr. Burke gave a clearly formulated picture of the conditions of preferential growth of a large grain in a fine grained matrix (Jeffries' grain size contrast effect) in terms of the Harker-Parker mechanism. This picture is entirely in accord with our own thoughts on the subject, provided the fine grained matrix has no preferred orientation. When this preferred orientation is not negligible, however, the picture needs to be supplemented by a mechanism which provides the strong dependence of the rate of growth of the large grain on its relative orientation with respect to its neighbors, as stated above. The exact nature of that mechanism is not yet completely clear.

References

1. Z. Jeffries: Grain Growth Phenomena in Metals. *Trans.* AIME (1916) **56,** 571.
2. Z. Jeffries and R. S. Archer: The Science of Metals. McGraw-Hill (1924).
29. O. Dahl and F. Pawlek: Kornordnung und Kornwachsturn bei Walzblechen. *Ztsch.* f. Metallkunde (1936) **28,** 266.
36. P. A. Beck, J. C. Kremer, L. J. Demer and M. L. Holzworth: Grain Growth in High Purity Aluminum and in an Aluminum-magnesium Alloy, AIME TP 2280. *Met. Tech.* Sept. 1947.
42. C. S. Smith: Grains, Phases and Interface. AIME TP 2387. *Met. Tech.* June 1948.
44. M. Cook and C. Macquarie: Development of Abnormally Large Grain Sizes in Rolled and Annealed Copper Sheet. *Trans.* AIME (1939) **133,** 142.
45. P. A. Beck, J. Towers, Jr. and P. R. Sperry; Grain Coarsening in Copper. *Trans.* AIME, **185,** 203: *Jnl. of Met.*, Feb. 1949.
46. J. S. Bowles and W. Boas: Effect of Crystal Arrangement on "Secondary Recrystallization" in Metals. *Jnl.* Inst. of Metals (1948) **74,** 501.
47. P. A. Beck and P. R. Sperry: Effect of Recrystallization Texture on Grain Growth. *Trans.* AIME, **185,** 240: *Jnl. of Met.*, March 1949.
48. H. Althertum: Zur Theorie der Rekristallisation. *Ztsch.* fur Metallkunde (1922) **14,** 417.
49. D. Harker and E. R. Parker: Grain Shape and Grain Growth, *Trans.* A.S.M. (1945) **34,** 156.
50. C. G. Dunn and F. Lionetti: The Effect of Orientation Difference on Grain Boundary Energies. *Trans.* AIME, **185,** 125: *Jnl. of Met.*, Feb. 1949.
51. C. G. Dunn: Controlled Grain Growth Applied to the Problem of Grain Boundary Energy Measurements. *Trans.* AIME, **185,** 72: *Jnl. of Met.*, Jan. 1949.
52. Weston Morrill: Improved Silicon-Irons for Electrical Equipment, *Metal Progress*, (1948) **54,** 675.

---

* University of Chicago

Statistical Rate Theory of Metals—I. Mechanism of Flow and Application to Tensile Properties

By JAY W. FREDRICKSON and HENRY EYRING

DISCUSSION

(G. Gensamer and W. R. Hibbard Jr., presiding)

G. WENSCH*—The authors have further advanced the basic fundamentals in the science of rheology. Upon the teachings of Bingham, Scott Blair, Houwink, and others we are approaching the solution of the inelastic behavior of metals.

In consideration of Fig 10, as the authors point out, the stress-strain curve follows line *MF* even after interruptions provided that relaxation effects do not occur. However if relaxation effects do occur, such that the behavior of the metal is inconsistent with the concept of an

---

* International Nickel Co. Graduate Fellow in Metallurgical Engineering, University of Illinois.

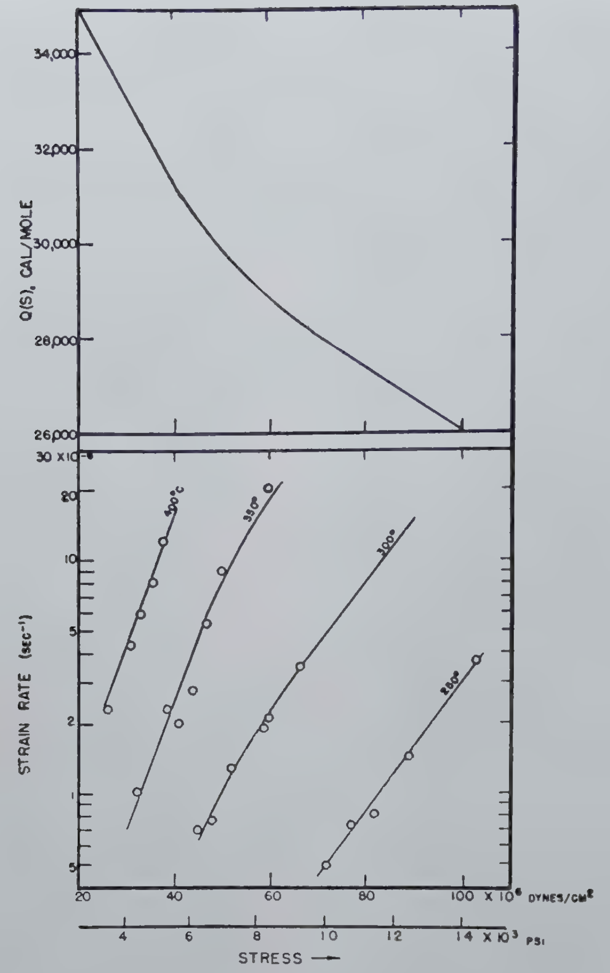


FIG 24—Analysis of Dushman's data after Ref. 43.

equation of state, transient phenomena will appear as clearly shown by the work of Carreker, Leschen and Lubahn.[42]

W. C. ELLIS*—The authors in this paper have advanced a theory for plastic flow in metals which involves the translation of relatively small flow units or domains into existing holes made up of a number of adjacent unoccupied lattice sites. From the mathematical analysis of the stress-strain curves for a number of steels they have calculated an average value for the hole involved in flow. A range of values was found which group around about 200Å^3 at room temperature. This volume corresponds to about 8 unit lattice cells which is a large hole in an atomic universe but a small one in relation to available measuring tools such as the electron microscope.

The mathematical analysis, by itself, seems to be merely suggestive and constitutes no proof of the physical existence of such holes, and it should be added that the authors do not contend that it does. Explicit evidence for the existence of such holes, if they be a reality, would be welcome for it would provide a firm foundation for a mechanism of plastic flow, a foundation which does not exist, in my opinion, for any of the present theories. I would therefore like to ask if there is a method, or prospect of developing methods, of arriving explicitly at the physical reality of such holes in metallic structure. This paper is a valuable one in that it presents a new concept and it is hoped that investigation along this line will be continued.

C. ZENER* and T. S. KÊ*—The authors conclude that the heat of activation ΔH^* for the elementary act of plastic flow is essentially zero, and hence, that the free energy of activation ΔF^* is entirely an entropy effect. This conclusion, if correct, must form the starting point for any further development of the theory of plastic deformation. It is of utmost importance that their underlying assumptions be carefully scrutinized to see if this conclusion of a zero ΔH^* could possibly arise simply as a consequence of an artificial assumption.

The authors assume that the rate of an elementary flow process depends upon the applied stress S in the manner:

rate $\sim e^{-(\Delta F^*/kT)}\{e^{(VS/kT)} - e^{-(VS/kT)}\}$, [42]

where the free energy of activation ΔF^* and the volume V may depend upon the temperature but not upon the applied stress. Now one of the most characteristic features of crystalline plasticity, as opposed to amorphous plasticity, is the relative insensitivity of the flow stress S to changes in temperature. Thus by passing from room temperature to 1°K the flow stress corresponding to a given strain rate is increased by, at most, a factor of five. The assumption (Eq 42) is compatible with this characteristic fracture of crystalline plasticity only if both the ratios V/T and $\Delta F^*/T$ are relatively insensitive to changes in temperature. A zero heat of activation ΔH^* then follows from the constancy of the ratio $\Delta F^*/T$. The conclusion of the authors that the free energy of activation is essentially an entropy effect is, therefore, a direct consequence of the assumption (Eq 42).

In writing Eq 42 one implicity assumes that the heat of activation Q of an elementary flow process is a linear function of the applied stress S. One of the writers (C. Z.)[43] has, in collaboration with J. H. Hollomon, shown that the extensive creep data of Dushman, Dunbar and Huthstein[18] may most readily be interpreted through a relaxation of this assumption, that is, by regarding the heat of activation as a function $Q(S)$, the precise form of which is to be determined by experiment. Thus, in place of Eq 42 one has

rate $\sim (e^{-Q(S)/kT} - e^{-Q-(S)/kT})$. [43]

The example of $Q(S)$ for pure aluminum is presented as Fig 24. Fisher and MacGregor,[28] whose data have been analyzed by Fredrickson and Eyring according to Eq 42, showed that their own data were consistent with Eq 43.

The present writers wish to ask Messrs. Fredrickson and Eyring if their physical picture of plastic flow necessarily demands a rate dependence containing the applied stress to the first power in the exponent, as in Eq 42, and whether a relaxation of this condition of an exponent linear in stress would not drastically modify their conclusions. They further wish to point out to Messrs. Fredrickson and Eyring that creep data at very low stress levels is not in agreement with the conclusion of a zero heat of activation. Such data have been analyzed by Kanter[44] who found that at those low stress levels where the creep rate is proportional to the stress itself, the creep rates for iron and for brass have heats of activation comparable to those for self diffusion.

J. W. FREDRICKSON and H. EYRING (authors' reply)—The authors wish to extend their appreciation for the interest shown in this paper.

The transient phenomena mentioned by Mr. Wensch has been observed[45] by the senior author in a subsequent investigation. A continuation of these transient investigations promises to enlighten us further on the flow mechanism.

---

* Bell Telephone Laboratories.

[42] References are at end of discussion.

* University of Chicago.

The question proposed by Mr. Ellis regarding the physical reality of "flow holes" in a metallic structure is a very intriguing problem and deserves serious consideration. As is emphasized by Mr. Ellis, the authors have not established the physical existence of such holes. It does seem to be a reasonable assumption however, that in order for flow to occur by the movement of domains these domains must be provided with a space to occupy in their new positions. A mechanism for flow has been explained by some theories through the operation of dislocations. It is not our wish to enter into a discussion at present as to the difference between a "dislocation," "imperfection" or "hole." It suffices to say that some medium is apparently required to explain flow and the medium of holes has been used in this paper to provide the mechanism for flow. If one assumes that the size of the domain that moves is an inverse function of the applied stress, as appears to be the case, then the mechanism of flow could also occur by the creation of holes or new "home sites" by the domain itself as it moves.

At present there is apparently no method of determining explicitly the physical existence of holes in metallic structures. As Mr. Ellis suggests, the development of such a method would contribute greatly to the knowledge of flow in metals. It is the authors' belief that such a determination can be secured by the methods of statistical mechanics and further work is being directed toward this end.

Messrs. Zener and Kê have presented a very interesting and thoughtful discussion. The points introduced in this discussion are well taken and merit consideration not only by the authors but by others interested in flow mechanisms.

The rate equation for the elementary flow process shown as Eq 42 in the discussion was developed after careful consideration of various flow processes occuring in solid materials including that of viscous flow. The fact that the apparent heat of activation ΔH^* for flow was essentially zero, and thus the free energy of activation ΔF^* was entirely an apparent entropy effect, was a direct consequence of the application of the elementary rate equations to metallic deformation. That this result might possibly have occurred by the use of some other assumption can not be categorically denied. But first, let us examine the equation a little more closely.

If in Eq 42 $VS \gg kT$ we have

$$\text{rate} \sim e^{-\frac{\Delta F^*}{RT}} e^{\frac{VS}{kT}} \qquad [44]$$

and the equation gives the same type of exponential dependence of rate on stress that is found empirically. On the other hand if $VS \ll kT$, one has approximate proportionality of the rate to the stress as in viscous liquids and as shown by Maxwell.[33]

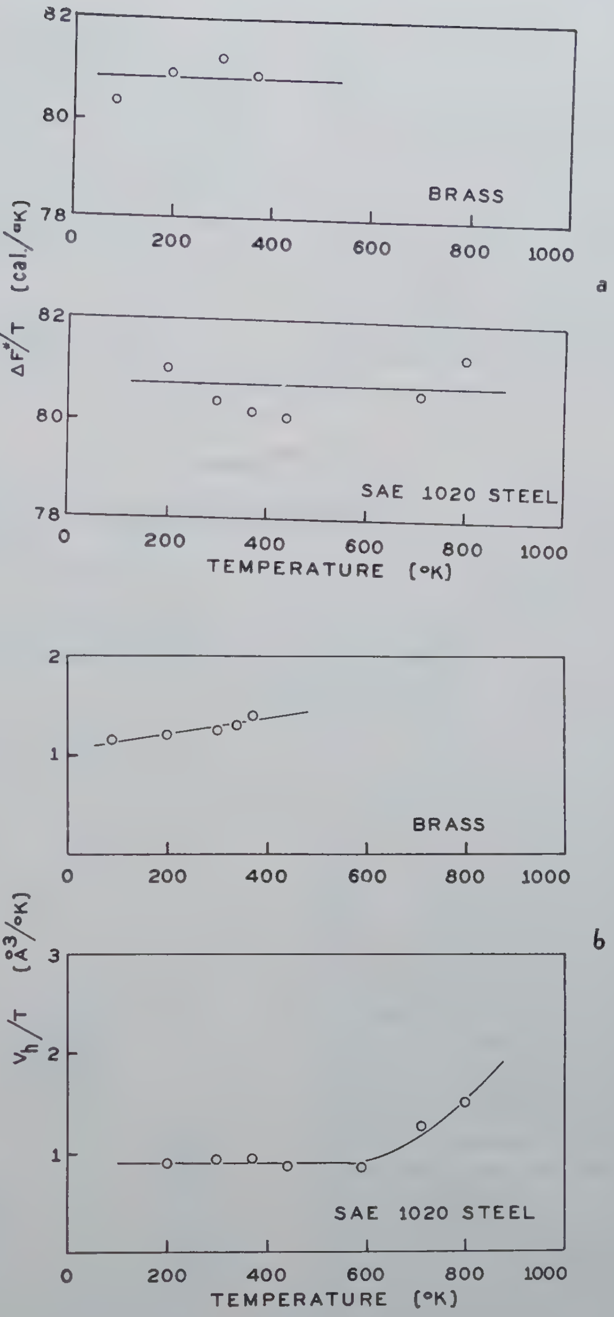


FIG 25—*(a)* *(above)* Curves showing $\Delta F^*/T$ vs. temperature relationship. *(b)* *(below)* Curves showing V_h/T vs. temperature relationship.

If one rewrites Eq 44 in the form

$$ln\ \text{rate} \sim -\frac{\Delta F^*}{RT} + S\frac{V}{kT} \qquad [45]$$

the stress is shown as a function of the logarithm of the rate. Eq 45 is clearly seen to fit the curves of Fig 24.[43]

A further analysis of the data in the paper shows that $\Delta F^*/T$ and V_h/T are relatively insensitive to changes in temperature, as was deduced by Messrs. Zener and Kê. Fig 25, representing Fig 2 and 3 from this analysis, shows the relationship of $\Delta F^*/T$ and V_h/T as a function of T.[46]

It has been determined by the senior author in a separate subsequent investigation of creep[45] that both the free energy of activation ΔF^* and the hole volume V_h are inversely dependent upon the applied stress. Because of this stress dependency, one must be careful in making an analysis that the underlying mechanism does not change within the stress range.

One may well consider that the mechanism of flow occurs by the process of self-diffusion in that at equilibrium the energy barriers that must be crossed are sufficiently high to prevent flow, or conversely, the probability for flow in any one direction is as likely as in any other and therefore flow will not occur without a directed driving force. The action of the applied stress is to provide the driving force for flow by raising and lowering the equilibrium positions on each side of the energy barrier as depicted by Fig 2. Thus, if the barriers are lowered in a preferred direction by some means, flow can occur by self-diffusion of domains across the barrier.

Messers. Zener and K have proposed a rate equation as shown by Eq 43. It appears that in order for this equation to fit the curves shown in Fig 24 the parameter Q must be some linear function of stress. If Q is some linear function of the stress, then Eq 42 and 43 differ only in the definitions of the exponents. Determining which equation is the more correct will require further investigation. It is the belief of the authors that the amount of data so far analyzed has been too meager to be used as a definite criterion to determine the nature of the parameters in the equation. In any event it is gratifying that separate investigations do produce such qualitative agreement.

References

42. R. P. Carreker, J. G. Leschen and J. D. Lubahn: Transient Plastic Deformation, AIME *Metals Tech.* Sept. 1948, TP 2477.

43. C. Zener and J. H. Hollomon: *Jnl. Appl. Physics* (1946) **17**, 69.

44. J. J. Kanter: *Trans.* AIME (1938) **131**, 385.

45. Unpublished report.

46. H. Eyring, J. W. Fredrickson and D. McLachlan, Jr.: The Mechanism of Flow for Solid Metals, *Proc.* Nat. Acad. of Sci. (1948) **34**, No. 6, 295–304.

Nucleation of Slip Bands

By J. G. LESCHEN, Junior Member, R. P. CARREKER, and J. H. HOLLOMON, Junior Member AIME

and

Transient Plastic Deformation

By R. P. CARREKER, J. G. LESCHEN, Junior Member AIME, and J. D. LUBAHN

JOINT DISCUSSION

(G. Gensamer and W. R. Hibbard, Jr., presiding)

G. WENSCH*—Messrs. Carreker, Leschen and Lubahn are to be congratulated on their excellent paper "Transient Plastic Deformation." As the authors correctly point out little thought has been given to transient phenomena in metallurgical experimentation; however, considerable effort has been made to analyze the rheological behavior of solids by mathematical treatment.

By the proper combination of visco-elastic models such as the Maxwell and Kelvin bodies one can imitate many of the observed behaviors in metals, such as the elastic after-effect, yield point and the transient behavior.[7,8]

With a slight modification of the visco-elastic model system given by Burgers[9] one can obtain a generalized strain-time curve similar to that of Fig 2 in the above-mentioned paper. By the proper evaluation of the constants of viscosity and elasticity the two curves would probably be unique. The writer submits a sketch of a visco-elastic model which would exhibit generalized transient behavior, see Fig 9.

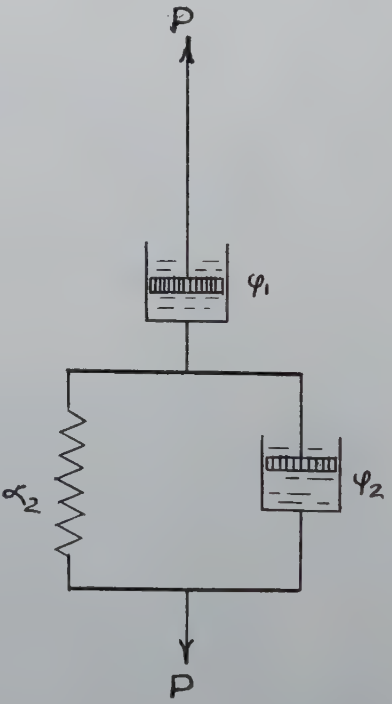


FIG 9—Rheological model which demonstrates transient behavior.

The solution of the differential equation of the model is:

1. $$x = \phi_1 \int_0^t P dt + \phi_2 \int_0^t P(t_1)\, e^{-(t-t_0)\phi_2/u}\, dt$$

2. when $P = P_1,\ 0 < t < t_1$

$$x = \phi_1\, tP_1 + \alpha\, 2P_1{}^{(1-e^{-t/u})}$$

$$\mu = \frac{\alpha_2}{\phi_2},\ P_1 > P_2$$

3. when $P = P_2\ t_1 < t < t_2$

$$x = \phi_1\, P_2(t - t_1) + \alpha_2 P_2\, (1 - e^{\frac{t_1 - t}{u}}) + \phi_1 P_1 t_1 + \alpha_2 P_1\, (1 - e^{\frac{-t_1}{u}})$$

By proper evaluation of the constants ϕ_1, ϕ_2 and α_2, the strain x may be plotted versus the time t to give a curve similar to the upper curve of Fig 2 in paper discussed. (Fig 10).

Perhaps the authors can be induced to comment on visco-elastic model analysis as a means of interpreting transient behavior.

It appears that the authors did not account for the change in the cross-sectional area of the wire which occurred during their investigation. Did the authors find this area reduction to be negligible?

J. G. LESCHEN (authors' reply)—With a suitable combination of springs and dash-pots one can undoubtedly reproduce many experimental results. The authors prefer not to depend too much on this sort of model since they feel that in many cases it does not represent adequately any real mechanism in the metal. However, this type of model is frequently of considerable value.

In the initial experiments no account was taken of the change in cross-sectional areas of the wires, since the reduction was negligible. In the experiments with which the sigmoidal curve was determined, the stress was automatically maintained constant by the device described by Fisher and Carreker.[10]

R. MADDIN* and W. R. HIBBARD, JR.*—The very interesting study of the dynamics of slip bands is a welcome addition towards the general solution of the problem of plastic deformation of metallic single crystals. There are two points which should be considered in light of recent experiments conducted at Hammond Laboratory.

In conjunction with Professor Henry A Lepper, Jr. (Department of Civil Engineering, Yale University), a single crystal of alpha brass 37 mm in diam was slit into 6 square crystals 180 mm long by 3.6, 4, 5.2, 7, 7.9 and 13 mm on the edge, respectively. Each crystal was etched to remove cold-work, homogenized for 20 hr at 800°C, polished electrolytically and X rayed via the back reflection method to insure that the machining operation left

---

* International Nickel Co. Graduate Fellow in Metallurgical Engineering, University of Illinois.

[7,8] References are at the end of discussion.

* Yale University.

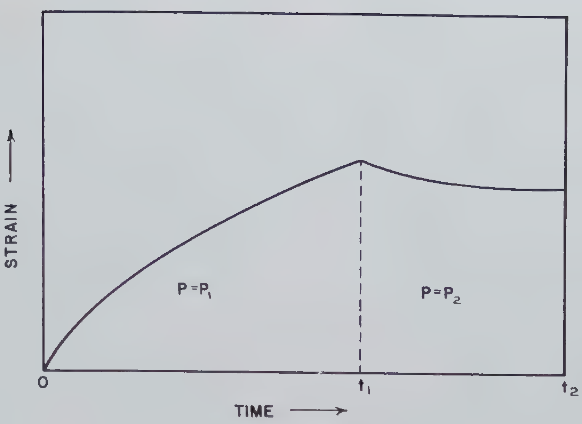


FIG 10—Schematic strain-time curve.

the orientation unchanged. An SR-4 strain gauge, type A-1, was glued on each specimen and the crystals were stressed to just beyond the yield point. The slip lines were counted on four of the specimens (too much glue on the remaining two specimens obscured the lines). The largest specimen showed a total of 70 lines observable at a magnification of 600 diam along a 150 mm gauge length whereas the smallest crystal showed over 8000 lines observable at 600× for a 25 mm (only ⅛ of gauge length). The number of lines on the other crystals was intermediate between those for the two extreme sizes and decreased with increasing size of the specimen. These preliminary results of this investigation, which is still in progress, have thus far produced experimental evidence which is not in accord with the statements made by the authors on p. 3 of their paper, "Nucleation of Slip Bands."

In the investigation of Ono quoted by the authors, two variables existed, namely, specimen diameter and orientation. Recent work being conducted at Hammond indicates that alpha brass single crystals of the same diameter pulled in tension under similar conditions but with differing orientations, when referred in the usual manner to the principal [111] plane in the principal <110> direction, give values for critical resolved shear stress too widely separated to be accounted for by experimental inaccuracies. Thus, variations in crystal orientations might account for Ono's differing critical resolved shear stress values. Furthermore, the curves redrawn from Ono in Fig 3 represent resolved shear stresses at shear equal to 4×10^{-5} and shear equal to 100×10^{-5} rather than a true critical resolved shear stress. The curve representing the resolved shear stress at shear equal to 0 (that is, the critical resolved shear stress) does not show so marked a downward trend and the data points deviate from the drawn curve making them difficult to generalize. It is also possible that the downward trend in Ono's curves and those redrawn in Fig 3 of the current paper might conceivably be rationalized on the basis of a variation of rate of strain hardening as a function of specimen diameter.

J. G. LESCHEN—We now realize our error in including in the present paper the curves from Ono and admit that they should be interpreted with caution.

However, a size effect is still to be expected. It should be a more complicated one than the present paper indicates, since the displacement per slip band δ and the number of slip bands N per unit length of specimen should also be considered. For example, if one assumes that the formation of a slip band in a specimen of diameter D relieves all the elastic energy stored in an approximate volume D^3, δ should vary as the product τD where τ is the stress at which the slip band started. The observed strain would then vary as $N\tau D$.

Now as I understand these experiments with brass crystals, the specimens were pulled to a constant strain and then the number of slip bands counted. Is that correct?

R. MADDIN—Relatively.

J. G. LESCHEN—Constant strain requires the product $N\tau D$ to be approximately constant. This requirement is consistent with the observations of Messrs. Maddin and Hibbard that N decreases with increasing D. It is difficult to say how τ should vary under these circumstances, and a report of any such variation in these experiments would be very interesting. The situation is further complicated by the likelihood that all the slip bands did not start at the same stress.

L. D. JAFFE*—The theory outlined by the authors is obviously worthy of most careful consideration, especially in view of its successful prediction of certain transient effects. I should like, however, to confine my remarks to Fig 4 of the paper, "Nucleation of Slip Bands," and the text pertaining to this figure.

The authors explain the portion of the observed yield strength vs. $T^{-\frac{1}{4}}$ curve that represents a straight line through the origin (Fig 4) on the basis of the first exponential in Eq 3. They attribute the deviation at higher temperatures to the influence of the second exponential and to decrease in the modulus G. However, the second exponent differs from the first simply by a factor of 2. A brief calculation shows that even if the whole variation in yield strength with temperature were attributed to the second exponential, the slope would be increased by a factor of $2^{\frac{1}{4}}$, or only 19 pct. The observed increase, according to Fig 4, is obviously very much greater than this. Thus, the second exponential of Eq 3 is of little help in accounting for the change in slope. Do the authors, therefore, believe that the deviation observed is primarily due to decrease in the modulus?

W. C. ELLIS*—Jay Fredrickson in our Laboratories has conducted creep tests similar to those for which Carreker, Leschen and Lubahn report the results in Fig 2 of their paper, "Transient Plastic Deformation." These tests were made on high purity lead with tensile specimens of rectangular cross-section, approximately 0.500 × 0.100 in. The strain rate curves as a function of time were similar to that shown in Fig 2 including the high rate of strain on initial application of stress decreasing slowly with time to a nearly constant rate. When the stress level was decreased the strain rate became zero for some time and then increased gradually to a nearly constant rate smaller than that for the initially higher stress. These are qualitatively the same transient effect as reported by the authors and would follow from the relationships developed in the companion paper, "Nucleation in Slip Bands."

The energy relationships which the experimental evidence qualitatively confirms were developed from the concept of nucleation in slip bands and the experimental findings are therefore qualitatively consistent with the nucleation concepts, as the authors have stated. There is a tendency when such an agreement is found between experimental data and a postulated mechanism to go further and to consider that the consistency observed points to a high probability for the physical reality of the mechanism.

There is a question, however, of whether a postulated mechanism, such as this, can be said to be unique from confirmation by property relationships alone, when such relationships might conceivably arise from other physical mechanisms. I would like, therefore, to ask Mr. Leschen to discuss this in more detail. How strongly do the experimental findings point to nucleation in slip bands as a part of the physical mechanism of flow in metals?

J. G. LESCHEN—Thank you, Dr. Ellis. I am glad to learn that someone else has performed this sort of experiment and obtained similar results.

---

* Watertown Arsenal Laboratory.

* Bell Telephone Laboratories.

I should like to emphasize that the present papers are not primarily concerned with the physical mechanism of slip. They simply seek to determine whether or not the nucleation concepts are applicable to the slip process. The results of the experiments indicate that slip bands do behave like anything else that is nucleated. Nucleation concepts should therefore be applicable to any mechanism of slip, whether that mechanism be described by dislocation theory or by the theory of Fredrickson and Eyring or by any other theory.

D. W. WHITE*—As I understand it, one corollary of this theory is that the small surface imperfections in the crystal would tend to facilitate the nucleation of the slip. On the other hand, Dr. McAdam showed us this morning that the low stress of a specimen on a larger scale is raised as you introduce the triaxial by means of a notch.

This seems on first appearance a little inconsistent with this matter of the function of surface imperfections by the nucleation theory. I was wondering if this is really a contradiction or just something imaginary on my part.

J. G. LESCHEN—The stress field in the vicinity of a notch is not as simple as that assumed in the present papers. It is difficult to say what the effect would be on the nucleation and growth of a slip band.

G. H. BOSS†—The authors of these two papers have outlined a very provocative theory and have presented the results of some very beautiful experiments. Their theory is certainly as tenable as are some of the others which attempt to explain slip. However, it is not necessary to postulate nuclei as the generators of slip in order to explain the curves which they have derived from the results of these excellent experiments.

While a metal is in the steady stage of creep it is being subjected to two opposing phenomena. One of these is the work hardening which occurs as the result of the deformation to which it is being subjected; the other is the recovery anneal, which occurs at a rate determined by the temperature to which the metal is heated.

In the following discussion it is assumed that the temperature is held constant, thereby fixing the rate of recovery annealing: If a metal is creeping at a steady rate caused by a load L, it will be work hardened to the extent determined by the balance between the rate of extension and the speed of recovery anneal. Now, if the load is increased by an amount delta L, the balance is upset because the metal has only the work hardening of the lower load to resist the newer, greater load. Consequently the metal is deformed at a very great rate, which gradually diminishes as the work hardening of the metal is increased. It finally reaches a steady rate which is somewhat higher than the original. Then if we reduce the load back to its original value of L the balance is again upset, because the metal is work hardened to the extent demanded by the higher load. Consequently the rate of extension is greatly reduced, and remains so until recovery has restored the balance by reducing the work hardening of the metal.

A. H. COTTRELL*—The idea that slip may be nucleated by the formation on a slip plane of a disc-like patch of slipped material which subsequently grows by the outward advancement of its boundary across the slip plane is much older than the authors suggest. It was proposed and discussed by Orowan[14] in 1934. In reintroducing the idea, the authors suggest that it is in some way different from the dislocation theory; this is strange in view of the fact that Orowan and many others have emphasized that the boundary (which divides the slipped and unslipped parts of the slip plane) of the disc-like patch constituting the nucleus of slip is rigorously definable as a dislocation. This model of slip nucleation has been discussed, explicitly and implicitly, many times[1,15,16,17,18,19,20] since 1934 and the general conclusion has now been reached that an extremely high energy of formation precludes the existence of slip nuclei as equilibrium features of a crystal under external stress, unless this stress approaches $G/10$ in order of magnitude, where G is the shear modulus. Accordingly, modern opinion holds that such nuclei exist only as metastable features in crystals, the dislocations which form their boundaries being trapped in potential energy troughs caused by the stress fields of impurities, crystal boundaries, and other dislocations. In the absence of these troughs, or of a very high shear stress on the glide planes, the nuclei must disappear.

The concept of a slip nucleus which the authors propose on p. 2 of "Nucleation of Slip Bands," as a basis for their theory is perplexing. It is founded on the assumption "that a growing semicircular interface, *across which no shear stress is supported*, advances slowly into the crystal." The clause in italics must mean that the material in the slip nucleus is incapable of supporting a shear stress, or that it happens to be free from stress at the time considered. From the work of Bragg,[18] who has studied the second possibility in detail, one can conclude:

1. That it arises when the applied shear stress τ bears a certain relation to the amount of slip (which is fixed by the crystal structure) in the slip nucleus, a relation which also involves the size of the nucleus and the elastic properties of the medium,

2. That, when its consequences are fully analyzed, it reduces the above model of slip to a particular case of the dislocation theory model, in which the stress fields from the applied load and the dislocation constituting the boundary of a slip nucleus approximately cancel each other inside that nucleus.

The authors evidently have in mind the first possibility (that is, that the slipped region cannot support a shear stress), since:

1. No suggestion is made of any critical relation between the applied stress and the amount of slip; in fact the amount of slip is not mentioned at all.

2. The energy of nucleation is taken to be made up of a term representing relief of shear stress in the vicinity of the nucleus and a surface energy term proportional to the area of the slipped region.

To suppose that the slipped region cannot support a shear stress must mean either that the material in the disc-like patch has become a fluid (see below) or that the two atomic planes which have slipped over each other in this region have parted company, thus forming a crack. The authors presumably favor the latter since they compare the disc-like patch to a slit.

What is now very difficult to understand is how this model can describe the process of plastic gliding since the picture it gives is of the spreading of a crack across a slip plane, which is something quite different. If the material within the slipped region cannot support a shear stress, then, when the "slip" has propagated across the entire slip plane, the two halves of the crystal divided by the slip plane must become separated. Brittle failure along a slip plane, and not plastic flow, should result.

The conclusion seems unavoidable that no other model of a slip nucleus is conceivable except the strict dislocation model in which the atoms in the slipped region have slipped past each other to congruent positions where they are as firmly bound and are as capable of supporting a shear stress as those in the unslipped region. Only the dislocation itself, where a special atomic configuration exists, is incapable of supporting a shear stress, and this only in the sense that the dislocation will move in that direction which causes the crystal as a whole to give way to the stress.

The fact that a crystal as a whole cannot support a shear stress without undergoing plastic flow, if it contains dislocations, raises the possibility that the authors may regard their slip nucleus as a macroscopic, disc-like piece of crystal containing many slip nuclei of the strict dislocation type. This is an alternative to the crack interpretation and could be regarded as a refinement of the concept of the slipped region as a "fluid." In such a case the energy required to form the macroscopic slip nucleus would be the sum of the self energies and interaction energies of all the nuclei (of the dislocation type) from which it is made up. We shall now show that the energy of formation is impossibly high for even a single slip nucleus of the dislocation type. Consider the formation of a circular patch of slip in the interior of a crystal, constituting a single slip nucleus of the dislocation type. (The slightly more complicated case of slip starting from the surface can also be treated by making use of Koehler's[17] surface image force.) The critical radius determining whether the nucleus will grow or disappear under a given stress is found by equating the forces acting on the dislocation forming the boundary of the nucleus. Three forces act:

---

* General Electric Co.

† U. S. Naval Base, Naval Air Material Center.

* University of Birmingham, England.

1. An outward force, causing the dislocation to spread, which is due to the applied stress. This force increases as the nucleus grows.
2. An inward force due to the attraction towards each other of the parts of the dislocation on opposite sides of the nucleus. This force decreases as the nucleus grows.
3. An inward force due to the fact that, as the nucleus grows, the length of the dislocation round its boundary increases and so the amount of highly strained material in the core of the dislocation increases. It can be shown[17,18] that the strain energy of the material in the core of a dislocation is generally smaller than the interaction energy of the strain fields which causes force (2), so that, in a rough calculation, force (3) can be neglected in comparison with force (2).

A balance between forces (1) and (2) is obtained approximately when the stress fields acting on any part of the dislocation due to the applied stress and the dislocation on the other side of the nucleus are equal. This defines the critical radius of the nucleus roughly as $\gamma = G\lambda/2\pi\tau$ where G and τ have their previous meanings and λ is the slip distance (~ 3Å). The energy of such a nucleus could be computed accurately, but for a rough estimate we can note that the energy of a dislocation is, at least, $1 eV$ per atom along its length.[20] Hence under a stress τ, the energy of nucleus formation is about $2\pi\tau/\lambda = G/\tau\ eV$. Strictly, one should subtract from this the work done by the external forces, but, unless these are such that $\tau > G/100$, the inclusion of this work does not affect the order of magnitude of the result.[19] For soft metals G is of the order of 1000τ, so that an energy of about 1000 eV is required for slip nucleation. With a Boltzmann factor of $e^{-40,000}$ operating, there can be no possibility of stable slip nuclei at the elastic limits of soft metals. Only under exceedingly high stresses ($\sim G/10$) could the possibility be admitted. While the above estimate is undoubtedly rough, it ought to be correct in order of magnitude. More exact treatments have been developed[17,18,19] for nuclei bounded by straight dislocations, which give essentially the same results.

The experiments on transient plastic deformation[7] cannot be regarded as providing evidence uniquely supporting the slip nucleation theory since Orowan[21] has shown that the same effects are to be expected on the recovery theory of creep. Indeed, the times of the induction periods noted ($\sim$ 10 min. for lead and $>$ 5 hr for copper) seem more in accord with a recovery theory than the theory of slip nucleation.

A. H. SULLY*—Although the paper, "Transient Plastic Deformation," makes no reference to the fact, it may be presumed that the wires on which these tests were carried out were polycrystalline aggregates and not single crystals. In this case, it seems more than probable that the results obtained in the tests, involving a reduction of stress during creep, are complicated by normal creep recovery effects. Creep recovery always accompanies stress reduction in polycrystalline aggregates, due to the redistribution of elastic stress between differently orientated crystals, which have different values of the resolved shear stress along the slip plane, and, consequently, have varying degrees of elastic and plastic distortion. In the case shown in Fig 2, where a small reduction of stress results in an apparently constant strain for a short period before creep recommences, it seems possible that this could be due to a balance between recovery processes in some crystals and continuing creep in others. If this were true, an actual diminution of strain would be expected to occur if the stress reduction was larger, and it is unfortunate that the authors do not give the results of experiments with larger reductions of stress, since this would be a more critical test of their theory. The statement in the introductory paragraphs that transient effects have not previously been recognized in experimental work is incorrect. Tapsell[22] has published data illustrating creep recovery and in particular has shown for a 3 pct nickel steel at 400°C the effect on the creep of reducing the stress from 16 tsi to 14 tsi. This shows, after allowing for the elastic strain, strain first diminishing, due to recovery effects, and then increasing and gradually approaching a new steady state creep rate.

The shape of the curve shown in Fig 3 from one minute onwards, relating the length of the period of zero strain rate to the time of prior application of the higher stress, can similarly be accounted for on existing theory if Andrade's division of creep into transient and quasi-viscous components is accepted and it is taken that only transient creep is recoverable in polycrystalline aggregates. There is adequate evidence that this is the case, for example, Chalmers[23] on the creep of tin. After the time of exhaustion of primary creep, the recovery accompanying a constant stress reduction therefore remains the same, irrespective of the period for which the higher stress is applied, in accordance with the authors' observed approach to constancy of their "incubation" period. The rapid increase in creep rate and subsequent decrease to a constant rate observed on increasing the stress can similarly be explained on the basis of existing theory by Orowan's[21] concept of an activation energy for glide increasing with the creep deformation along a curve similar to the normal stress-strain curve. An increment in the stress therefore results in a sudden lowering of the activation energy for further slip and consequently a further stage of transient creep, the exhaustion of which corresponds with the achievement of a constant creep rate, observed by the authors. It may be suggested, therefore, that the authors' results can be satisfactorily explained on the basis of existing knowledge of transient creep, and of the deformation of polycrystalline aggregates, and the evidence in favor of the theory of Leschen, Carreker, and Holloman would only be convincing if they observed identical effects, especially those accompanying stress reduction, in single metallic crystals.

J. G. LESCHEN—The authors regret that a misunderstanding has arisen with Professor Cottrell through their apparent failure to make sufficiently clear certain points in their introduction. They did not mean to imply that the idea of an advancing boundary of slip was at all new or that the dislocation and nucleation concepts are incompatible with one another. The authors merely wished to suggest that since nucleation theory can be successfully applied to the growth of cracks and of particles of a second phase, it might likewise be applicable to the growth of slip bands. In order to determine if it were so applicable, they selected a simple model—undoubtedly an over-simplified one—and then qualitatively investigated the characteristics of that model. Since these characteristics proved to be in reasonably good agreement with experiment, as reported both in the present papers and in a later development by Hollomon,[24] a nucleation theory of slip seems to merit further attention and a more quantitative development.

The present rudimentary treatment did not seem to warrant any undue concern with the atomistic mechanism by which slip takes place. In future developments of the theory the mechanism will have to be taken into account, and there can be little doubt that dislocations will here play a large part. In this regard the comments of Professor Cottrell should be of great help. The authors take issue with several of his remarks but feel that a discussion of most of them belongs more properly in a future note or paper

---

* Fulmer Research Institute, England.

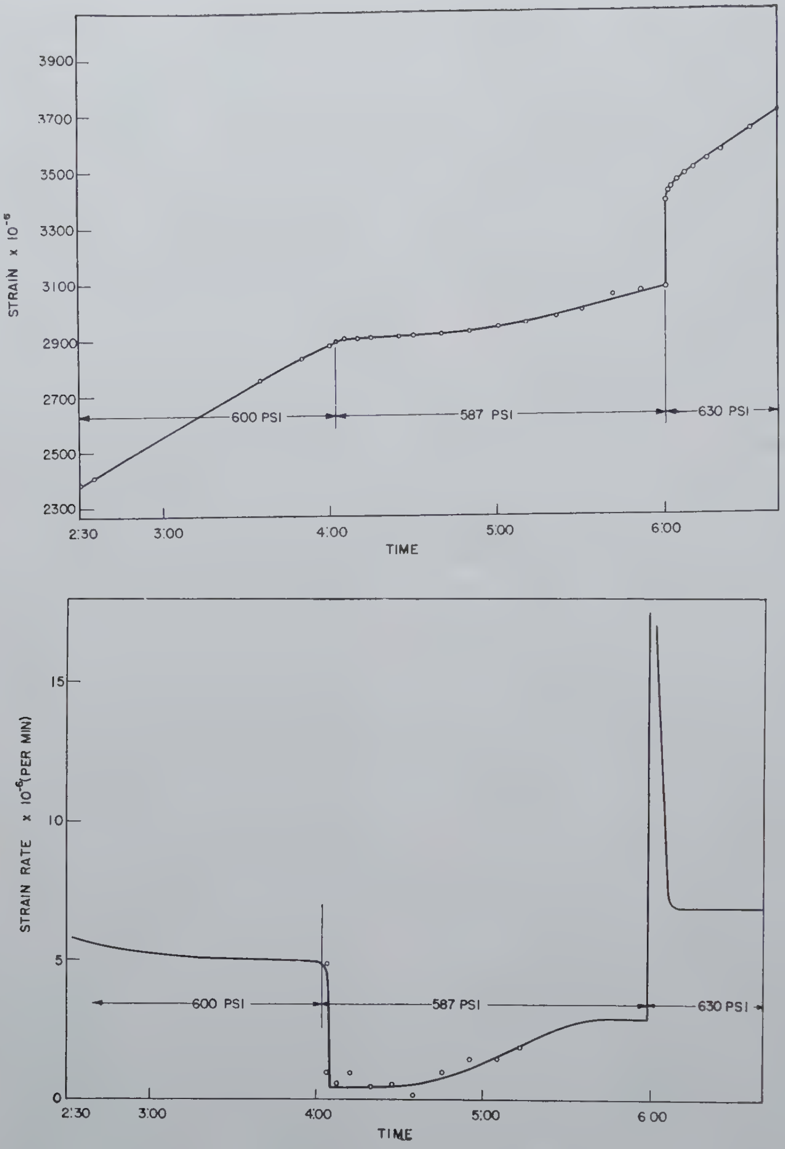


FIG 11—Transient effects in creep test on single crystal of lead.

than in the present discussion. However, several points can be made. For example, in calculating the Boltzmann factor for nucleating a dislocation, Professor Cottrell neglects the fact that yielding may begin at places where the local stress is significantly higher than the average applied stress. At such places, as Cottrell points out, the work done by the external forces cannot be neglected. Furthermore, we wonder whether the value of 1 eV, taken from the calculations of Nabarro, is a total energy or is it the free energy that must be used in computing the Boltzmann probability.

Mr. Sully is quite correct in presuming that the wires used in the original experiments were polycrystalline. However, tests have since been carried out with single crystals, and the results of the few experiments thus far completed indicate that the behavior of single crystals is essentially the same as that of polycrystalline specimens. Fig 11* illustrates, for instance, the response of a lead specimen to changes of stress at room temperature. Etching and visual examination of this particular specimen before the creep test indicated that it was a single crystal. X ray diffraction photographs made several days after the test showed the specimen to possess some lineage structure and a few small grains (possibly from local recrystallization). The suggestion that experiments be made with larger reductions of stress is a very valuable one and the authors have undertaken such tests.

Although the paper by Tapsell is not at present available to the authors, the type of results which he observes, as reported by Mr. Sully, is familiar.[25] However, this sort of transient does not appear to be the same as that discussed in the present papers, since the authors observed no contraction other than an immediate elastic one upon reduction of the stress. The transients of Tapsell probably require more complicated explanations of the kinds summarized by Leschen[26] and Zener.[27]

In reply to Dr. Jaffe, it does indeed appear that the decrease of modulus with increase of temperature is primarily responsible for the deviation of the yield strength versus $T^{-1/4}$ curve from a straight line. The work of Kê[28] indicates that the decrease in modulus of single crystals is appreciable; and there is reason to believe[29] that as the melting point is approached the modulus, of polycrystals at least, falls toward zero.

Messrs. Cottrell, Sully, and Boss point out that the observed transient effects can be explained equally well by the creep recovery theory of Orowan. An experiment which would differentiate between this and the nucleation theory is difficult to conceive. A choice between the two will perhaps depend upon which can be the more adequately developed.

The nucleation theory of slip will no doubt require considerable modification. Meanwhile, it should serve to direct attention to the type of experiment involving sudden changes of stress, strain rate, or temperature. Then, if the theory should ultimately prove to be untenable, it will at least have been of heuristic value.

---

* The strain rate-time curve was obtained by graphically determining the slope of the strain-time curve, except in the region following the decrease of stress. Here the strain-time curve was so flat that accurate measurements of the slope could not easily be made, and the method of differences was therefore used. This method yielded points with appreciable scatter, and it accordingly seemed appropriate to show them in Fig 11.

References

(Transient Plastic Deformation)

7. R. Houwink and W. G. Burgers: Elasticity, Plasticity and Structure of Matter, Cambridge (1937), 64.
8. A. M. Freudenthal: Notes on The Inelastic Behavior of Materials and Structures, Department of Theoretical and Applied Mechanics, University of Illinois, Urbana (1947) Parts I and II.
9. J. M. Burgers: First Report on Viscosity and Plasticity. Amsterdam (1935), Chap. II.
10. J. C. Fisher and R. P. Carreker: A Simple Constant Stress Creep Test. *Trans.* AIME, **185,** 178: *Jnl. of Met.*, Feb. 1949.

References

(Nucleation of Slip Bands)

1. F. Seitz and T. A. Read: *Jnl. Applied Phys.* (1941), **12,** 100, 170, 470, 538.
7. R. P. Carreker, J. G. Leschen and J. D. Lubahn: AIME, *Metals Tech.*, Sept. 1948, TP 2477.
14. E. Orowan: *Ztsch.* f. Phys. (1934), **89,** 634.

15. J. M. Burgers: *Proc.* Roy. Acad. Sci. Amsterdam (1939), **42,** 293; *Proc.* Phys. Soc. (1940), **52,** 23.
16. E. Orowan: *Proc.* Phys. Soc. (1940), **52,** 8; Inst. Met. Symp. on Internal Stresses (1947), 47.
17. J. S. Koehler: *Phys. Rev.* (1941), **60,** 397.
18. W. L. Bragg: *Nature* (1942), **149,** 511; *Trans.* N.E. Coast Inst. Eng. Shipbuilders (1945), **62,** 25; Inst. Met. Symp. on Internal Stresses (1947), 221; Report of Bristol Conf., Phys. Soc. (1948), 26.
19. F. R. N. Nabarro: *Proc.* Phys. Soc. (1947), **59,** 256.
20. N. F. Mott and F. R. N. Nabarro: Report of Bristol Conf., Phys. Soc. (1948), 1.
21. E. Orowan: *Jnl.* West Scotland Iron and Steel Inst., 1947.
22. H. J. Tapsell: *De Ingenieur*, 1946, **30,** (2), 57.
23. B. Chalmers: *Jnl.* Inst. Metals, 1937, **61,** 103.
24. J. H. Hollomon: Nucleation of Deformation. ASM Symposium on Cold Working of Metals (1948).
25. J. D. Lubahn: Creep of Metals. Ibid.
26. J. G. Leschen: Plastic After-effects. Ibid.
27. C. Zener: Anelasticity of Metals. TP 1992, *Metals Tech.* (Aug. 1946): *Trans.* AIME (1946) **167,** 155.
28. T. S. Kê: Experimental Evidence of the Viscous Behavior of Grain Boundaries in Metals. *Phys. Rev.* (1947) **71,** 533.
29. L. Brillouin: On Thermal Dependence of Elasticity in Solids. *Phys. Rev.*, (1938) **54,** 916.

Plastic Deformation of Large Grained Copper Specimens

By W. R. HIBBARD, JR., Junior Member AIME

DISCUSSION

(R. F. Mehl and J. C. McDonald presiding)

R. F. MEHL*—It is certainly pleasant to have a paper in the field of the plasticity of metal crystals that modestly undertakes to study some of the complexities from a phenomenological point of view, advancing new and good information and employing theory with judgment and restraint. In the long view, progress ought to come from a nicely calculated mixture of theory and experiment; and the proportions of each in that mixture should be chosen with some regard for the stage of development and the degree of complexity involved. Despite the great attention that has recently been given to physical theory in the field of metal behavior generally—the many symposia, the innumerable papers, the approval that is accorded semi-religiously to any effort so long as dislocations or lattice vacancies are intoned—it seems clear to this observer that progress in understanding plasticity has been shockingly slight. A proper sense of responsibility should be shown in a field of research; theory should not far outrun experiment, for even though there is good sport in this, the purpose is not one of entertainment. Nor should undirected experiment be the rule; but at the present the error is not in this. There is a need of systematic experimentation in this field, and from this a formulation of general modes of behavior, for which then we might hope to have theories of behavior, and not before. The metallurgical profession ought to remember this, as Dr. Hibbard evidently does, and as some physicists do not.

As to the problem at hand, the plasticity of aggregates, it is possible that some of the difficulties arise in a lack of knowledge of the conditions of flow under various circumstances of restraint. We have based our thinking on the Polanyi-Schmid principle of a critical resolved shear stress for slip, providing slip on that lattice plane and in that direction for which the resolved shear stress is a maximum; it is possible that if this principle is not general, we could profitably restudy aggregates. Dr. R. Smoluchowski in the Carnegie Metals Laboratories has recently demonstrated that slip in single crystals of aluminum in the form of thin plates does not conform to the Polanyi-Schmid principle, but that the length of the slip direction through the piece itself is a determinant—that slip is the readier the shorter this direction, that slip may occur in a given plane in a given direction when the length of the slip direction is short even though this slip system does not exhibit the maximum resolved shear stress. This effect of external geometry is in the category of restraint; it would be very interesting to see whether the change in basic thinking which this experiment implies might alter our conceptions of how aggregates should behave.

W. R. HIBBARD, JR. (author's reply)—I should like to thank Dr. Mehl for his kind remarks. His philosophy of scientific progress is a very important guide to investigations and, unlike the chicken or egg story, gives us a lead to the answer of which should come first, theory or experimental fact.

Dr. Smoluchowski's findings are extremely interesting and important. We now have another variable to consider in deformation studies.

---

* Carnegie Institute of Technology.

Plastic Flow in Anisotropic Sheet Steel

By L. R. JACKSON, Member AIME, K. F. SMITH and W. T. LANKFORD

DISCUSSION

(R. F. Mehl and J. C. McDonald presiding)

R. HILL*—Apart from notation, the anistropic stress-strain relations used by the authors are equivalent, in the special case when the directions of principal stress and orthotropy coincide, to those which I have derived by other methods in an earlier paper.[7] The criterion of yielding which I proposed is also similar in form to the authors' expression $\bar{\sigma}$, although I cannot make out whether they actually identify $\bar{\sigma}$ with the function governing yielding. It seems that they do not, since in the second paragraph of their summary and conclusions they state that their method is consistent with the idea that "the flow is governed primarily by considerations of shear-strain energy."

The value of the paper would have been greatly enhanced if detailed data had been given concerning the observed strain-ratios in the bulge tests. These can be predicted theoretically from the stress-strain relations in terms of the measured strain-ratios in the tension tests at right-angles. An immediate test of the validity of the stress-strain relations is then possible. Would the authors be good enough to give these data? I should also be very interested to check the relations given in my paper between the yield stresses in the orthotropic directions and the parameters occurring in the stress-strain relations.

---

* Cavendish Laboratory, Cambridge, England.

[7] A Theory of the Yielding and Plastic Flow of Anisotropic Metals. *Proc.* Roy. Soc., (1948) 193-A, 281.

I have found it very difficult to make out what fundamental postulates the authors are making with regard to the influence of cold work on the values of the anisotropic parameters. They appear to assume tacitly that these parameters increase in strict proportion. This seems rather unlikely on general grounds, but if it is true it certainly makes the problem of plastic anisotropy considerably easier. (In my own paper I left open the question of the dependence of the anisotropic parameter on the strain-history.) The second postulate made by the authors appears to be that the effective stress $\bar{\sigma}$ is a function only of the total plastic work. This is the natural assumption to make by analogy with the isotropic theory. However, could it not have been tested more directly by comparing the results of the two tension tests at right-angles? Would the authors please give the relevant data for this comparison?

The agreement between the bulge and tensile tests is, on the face of it, very satisfactory, but here again the authors have withheld the data that would enable a reader to assess whether the agreement is significant.

R. F. MEHL*—I would like to ask Mr. Jackson what are his ideas concerning the departure of Cor-Ten in behavior from his calculations.

L. R. JACKSON (authors' reply)—We have no concrete ideas as to just why Cor-Ten behaved as it did; however, throughout the analysis we made assumptions which may not have been justified in the case of Cor-Ten. Three such assumptions are: 1. That the flow curves in tension and compression are identical. 2. That the materials being investigated were orthotropic and that the orthotropic axes could be identified with the length, width, and thickness directions in the sheet. 3. That differences in the rate of straining in the bulge and tension tests had a negligible effect on the shape of the flow curves. 4. That the bulge test applies stresses only along the orthotropic axes. This last assumption is not correct and is particularly evident for circular bulges where there is no real choice of axes. 5. We also believe that the methods of measuring the parameters K_{xs} and K_{ys} are crude and not particularly precise.

R. F. MEHL—Do you think it possible that your results might be affected by variations in the ductility properties in the thickness direction?

L. R. JACKSON—We attempted to minimize the effect of nonuniform thickness by measuring the thickness of test pieces at numerous locations and selecting only those which were uniform within narrow limits.

R. F. MEHL—It is disturbing that we have hitherto been forced to consider plastic properties, even when the behavior in the transverse direction is important, on the assumption that the ductility in the transverse direction (the thickness direction in plate or sheet) shows no special characteristics. Recent studies at Carnegie on the transverse mechanical properties of heat-treated forging steels show that the transverse ductility properties vary greatly from the longitudinal; and though this doubtless originates in the structural anisotropy caused by alloy segregation and its consequences in heat-treatment, it is possible that a similar mechanical anisotropy, perhaps less in degree, may occur in simpler alloys; there is no information on this point, and it should be developed. The effect might well be present in Cor-Ten, which the author studies, for this is an alloy steel, and should show a degree—not necessarily a harmful one—of alloy segregation and resultant mechanical anisotropy.

M. F. SAYRE*—Was any measurement made on stresses in the 45° direction? From the theoretical point of view, evidence of anisotropy should be much more evident with measurements at 45° than with measurements taken 90° apart.

L. F. JACKSON—We did not measure the tensile properties in directions other than the rolling direction and the width direction; however, we are making these tests now.

MEMBER—How did you calculate the stresses in the elliptical bulge?

L. F. JACKSON—We had some difficulty in this measurement. We did not believe that the conventional method of computing stresses by assuming that the ellipse is a figure of revolution was precise enough, so the method had the following steps: 1. A bulge was blown to some selected height and the pressure released. 2. Either Tuckerman or wire strain gauges were mounted at the top of the bulge along the rolling direction and width direction. 3. The oil pressure under the bulge was then gradually increased to the value it had when the pressure was released. Simultaneous strain and pressure readings were taken during this period and a relation between the two strains and the pressure was obtained. The stresses existing at the maximum pressure were then computed from the strains by conventional methods.

MEMBER—How constant did that ratio remain during your test?

L. F. JACKSON—Do you mean the ratio between the stresses in the longi-

Table 3 . . . Bulge Stress-strain Data

Test and Material	Stress in K psi				R_x Inches	R_y
	σ_x	e_x	σ_y	e_y		
Rimmed, annealed, long. elliptical (Fig 4)	52.8	0.0640	67.0	0.0956	16.0	7.7
	55.6	0.0777	69.4	0.1196	14.1	7.1
	60.0	0.0934	71.4	0.1556	12.5	6.7
Rimmed, annealed, trans. elliptical (Fig 5)	56.2	0.0745	47.2	0.0409	8.8	18.4
	61.0	0.1024	52.8	0.0546	7.8	16.1
	65.1	0.1239	56.6	0.0654	7.3	14.3
Rimmed, annealed, circular, (Fig 6)	51.5	0.0506	50.9	0.0494	12.9	12.5
	56.6	0.0697	55.3	0.0690	11.0	10.9
					9.1	9.1
	64.2	0.1096	62.7	0.1105		
Al. killed, annealed long. elliptical (Fig 7)	44.4	0.0321	58.8	0.0505	22.5	10.6
	48.3	0.0393	63.0	0.0638	19.9	9.3
	49.6	0.0458	66.2	0.0768	18.2	8.7
	52.7	0.0540	68.9	0.0944	16.5	8.0
Al. killed, annealed transverse elliptical (Fig 8)	56.1	0.0519	53.8	0.0334	10.5	20.0
	61.0	0.0701	54.8	0.0440	8.5	17.5
	64.1	0.0850	57.1	0.0522	7.9	16.5
	67.6	0.1028	61.2	0.0620	7.3	14.8
Al. killed, annealed circular (Fig 9)	48.5	0.0420	48.5	0.0344	13.3	13.5
	53.8	0.0524	55.8	0.0442	12.0	12.1
	55.0	0.0623	57.4	0.0538	11.1	11.2
	57.5	0.0722	61.0	0.0637	10.3	10.3
	58.9	0.0853	62.3	0.0766	9.7	9.7
Al. killed, T.R. long., elliptical (Fig 10)	45.5	0.0327	56.0	0.0521	20.5	9.9
	49.1	0.0449	60.6	0.0736	17.6	8.6
	53.3	0.0574	63.9	0.0983	15.5	7.7
	56.6	0.0699	67.3	0.1193	13.1	7.1
Al. killed, T.R. trans. elliptical (Fig 11)	55.8	0.0529	44.7	0.0315	9.9	20.6
	60.8	0.0732	50.0	0.0422	8.6	17.8
	64.6	0.0994	54.2	0.0564	7.6	15.6
	67.0	0.1174	55.7	0.0659	7.2	14.3
Al. killed, T.R. circular (Fig 12)	54.3	0.0499	52.9	0.0484	12.8	12.8
	58.0	0.0663	56.5	0.0638	11.1	11.0
	60.6	0.0810	59.3	0.0782	10.2	10.3
	64.8	0.0991	62.5	0.0955	9.5	9.6
Cor-Ten long, elliptical (Fig 13)	76.0	0.0282	95.3	0.0404	22.5	11.3
	80.8	0.0367	101.6	0.0528	20.2	9.9
	86.0	0.0504	113.1	0.0748	16.9	8.6
Cor-Ten trans. elliptical (Fig 14)	84.2	0.0249	70.1	0.0158	13.2	27.3
	92.5	0.0366	71.6	0.0230	11.2	22.9
	100.5	0.0479	76.7	0.0302	9.9	20.4
	104.3	0.0678	86.1	0.0420	8.7	17.6
Cor-Ten circular (Fig 15)	93.9	0.0576	104.0	0.0536	12.4	12.4
	97.3	0.0674	100.5	0.0636	11.6	11.6

---

* Carnegie Institute of Technology.

* Union College.

tudinal and transverse directions?

MEMBER—Yes.

L. F. JACKSON—The ratio remained quite constant for any one test; however, the ratio obtained depended on the kind of steel.

The authors regret that Dr. Hill's excellent paper was not available at the time theirs was submitted and are glad to have the opportunity afforded by his discussion to acknowledge it.

With regard to his suggestions we have the following comments.

Within the precision of our measurements which was not high, the anisotropy parameters do not vary with cold work within the range covered. We wish to point out, however, that we did not explore the region in the vicinity of the yield point in detail because the assumption of incompressibility is not valid there. The parameters appeared to be essentially constant even after necking started when measured at the minimum section of the neck.

We are inserting Table 3 giving the data on the bulge tests requested by Dr. Hill. In the table we are also giving the radii of curvature at the top of the bulge, since this data would be required in order to take into account the fact that the bulge tests do not apply stresses only along the orthotropic axes. This should be taken into account in attempting to compute the strain ratios as Dr. Hill suggests. As noted in the paper, we did not take this factor into account in computing $\bar{\sigma}$.

The tension data for the two directions can be computed from the tension curves given in the figures through the use of Eq 21 and 22.

The Effect of Thermal-mechanical History on the Strain Hardening of Metals

By J. E. DORN, Member AIME, A. GOLDBERG and T. E. TIETZ

DISCUSSION

(R. F. Mehl and J. C. McDonald presiding)

G. W. GEIL[*] and N. L. CARWILE[*]—Tests of a type similar to those reported in this paper were carried out recently by the writers at the National Bureau of Standards with ingot iron. Complete true stress-true strain curves, up to the point representing fracture, were obtained for specimens prestrained at one temperature and extended to fracture at a different temperature. The curves were derived from simultaneous load and diameter measurements made during the tension tests. The results are in agreement with those reported in this paper in that they do not conform to the assumptions proposed for a mechanical equation of state. Tension tests made previously with an ingot iron, an 0.12 pct carbon steel and an 0.46 pct carbon steel, in which the specimens were prestrained various amounts at room temperature and then extended to fracture at 85°K gave similar results. These results were discussed in a paper on the "Influence of Plastic Deformation, Combined Stresses and Low Temperature on the Breaking Stress of Ferritic Steels" by D. J. McAdam, Jr., G. W. Geil and R. W. Mebs, published in *Metals Technology*, Aug. 1947.

The results of tension tests made at room and at low temperatures on commercially pure aluminum (99.4 pct) and high purity aluminum (99.97 pct) were discussed in a paper on the "Effect of Combined Stresses and Low Temperatures on the Mechanical Properties of Some Non-ferrous Metals" by D. J. McAdam, Jr., G. W. Geil, and R. W. Mebs published in the *Transactions* of the American Society for Metals, (1946) 37. The true stress-strain curve for aluminum extended at 85°K, warmed to room temperature, and subsequently extended further at 85°K, was not a continuous curve. The warming of the metal to room temperature resulted in some softening or crystal recovery. On subsequent retesting at 85°K, the stress-strain curve was lower than that which would have been obtained if the specimen had not been warmed to room temperature. The softening of aluminum prestrained at 85°K, warmed to 194°K, and retested at 85°K should be less than the softening resulting from warming to room temperature. This softening effect also is indicated by the curves in Fig 12 and 13 for aluminum as presented by Dorn, Goldberg and Tietz. Fig 12 shows the effects of prestraining at 78°K on the true stress-true strain curves at 292°K. The slope of these curves is due to the combined influence of the ordinary strain hardening at room temperature and the room temperature softening following the previous strain hardening at the lower temperature.

The slope of these curves decreases as the amount of prestrain at 78°K increases. With extension to maximum load or necking these curves approach the curve obtained in the single stage test at 292°K. Fig 13 presents the effects of prestraining at 78°K on the true stress-true strain curves at 194°K. With extension to necking, however, these curves still remain considerably above the curve obtained in the single stage test at 194°K, indicating less softening than in the test results shown in Fig 12.

J. H. HOLLOMON[*]—Inherent to the problem of plastic deformation is the relationship between the effects of strain introduced under differing conditions. In treating the engineering problem of materials, the simplest assumption that can be made with respect to prior strain history is that its effect is negligible if no phase transformations take place during the deformation. This assumption was made in the early part of this century by Ludwik and was justified by a few experiments. Other investigators have utilized the qualitative conclusions derived from this assumption in discussing and describing the effects of temperature and strain rate on the resistance to plastic flow. Certainly the notion permits a rationalization of such gross features of deformation as the relations between necking in tension and in creep tests.

On the basis of a few experiments on steels, I showed that the effects of prior strain history were exceedingly small. More recently, experiments have been performed by G. Sachs and E. J. Ripling[*] that are in almost complete agreement with my earlier experiments. It appears that particularly at low temperatures, and for ferritic steels, there is only a small effect of prior strain history even at large strains. On the other hand, the results of the authors show a pronounced effect of prior strain history. Orowan[†] has shown that the effect of prior strain history becomes progressively greater with increasing strain. The authors of the present paper have carried out a thoroughly competent investigation of the effect of strain history for their materials.

In Fig 18 of their paper, the authors show a difference in the equivalent strain depending on whether or not the specimen was first deformed at a high or a low temperature. Another equivalent strain may be definable that is independent of the sequence of the deformation temperatures. It seems reasonable that the shift in the stress strain curves should have been such as to make them assympotic at large strains rather than identical at smaller strains. If the coincidence was assured at large strains, then the equivalent strains for the two sequences of temperature changes would have been nearly identical. There is some basis for defining an equivalent strain in this fashion. One would expect transient effects immediately upon reapplication of the load at the new temperature which would disappear with deformation at the second temperature.

---

[*] National Bureau of Standards.

[*] General Electric Co.

[*] The Effect of Strain-temperature History on the Flow and Fracture Characteristics of an Annealed Steel. *Trans.* AIME Vol. **185**, 78, *Journal of Metals*, Feb. 1949, 78.

[†] West of Scotland Iron and Steel Inst. (1947).

Aside from such transient effects, the authors' experiments by themselves in no way invalidate the concept of the mechanical equation of state. The authors have simply shown either that there is no mechanical equation of state or that plastic strain is not an adequate measure of deformation. The self consistency of their results and an intuitive feeling with regard to the behavior of matter seem to require the latter conclusion. In order to demonstrate definitely that there is in fact little or no effect of prior deformation history when deformation is properly defined the following kind of experiment is necessary. Two specimens that are deformed the same amount (to the same equivalent strain in the authors terminology) must be deformed at some third temperature and their flow curves determined. If then these flow curves are identical, then the existence of a mechanical equation of state would have been demonstrated. Such a critical experiment is important to the further rationalization of the mechanism of deformation. It is conceivable that the equivalent strain if it does exist measures some fundamental properties of the material. Dr. D. J. McAdam,* and Dr. J. C. Fisher and I† have suggested that perhaps a proper measure of deformation fulfilling this condition is the stored internal energy. At a single temperature stored energy is probably a monotonic function of strain. However, this functional relationship may be different at different temperatures. Under such conditions, the work done during deformation may be a better measure of the stored energy. When the stress data of these investigators and of Orowan are plotted versus work, much less effect of prior strain history is noticeable.

Certainly the concept of equivalent strains is a good approximation to the way the authors' specimens behaved. If such equivalent strains are determined properly as can be done by investigations similar to the authors, and the significance of these equivalent strains is rationalized, a long step forward in understanding plastic deformation and predicting creep behavior would have been made. I fully agree with the authors that their data imply that at least for large strains all previous relations between the strength of metals and temperature are utterly confused by including both a variation in strength itself and a variation of the degree of deformation.

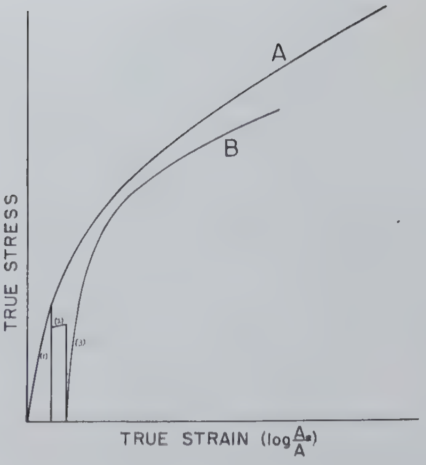


FIG 20—Effect of thermal prior strain history.

D. H. WOODWARD*—I would like to mention some results obtained at the National Bureau of Standards on the effect of thermal prior strain history. Curve *A* in Fig 20 was the type obtained at room temperature with the annealed material and curve *B* was the type obtained with the prior strained material. The schedule of prior straining was as follows: 1. The annealed material was extended 32 pct at room temperature. 2. The load was removed, the temperature raised, and the specimen deformed a small amount at a creep rate. 3. The load was removed, the temperature lowered to room temperature, and the flow curve obtained at that temperature. In such a case, not only is the shape of the latter flow curve different, that is, it is impossible to transpose it, but the resultant ductility is reduced.

J. H. HOLLOMON—That is 35 pct at room temperature, then heat it to what?

D. H. WOODWARD—Heat it to 850°F, that is, room temperature deformation followed by heating to 850°F. Deform a small amount in creep. Then complete the test at room temperature. The material was Monel.

E. J. RIPLING†—I should like to add here that there is no reason to suppose that results obtained on the fracturing characteristics after straining at two different temperatures are necessarily related to the flow characteristics under these conditions.

For example, work has been done at Case in which steel specimens were first prestrained at room temperature and then tested at some lower temperature. These results, which should be published shortly, indicate the very complex behavior of the fracturing characteristics in a two-step test. When the retained ductility at the low temperature was plotted as a function of the prestrain conducted at room temperature, the curve possessed a very sharp minimum and maximum. However, the flow stress for these same specimens could be represented by a smooth curve always possessing a positive shape.

D. J. McADAM, JR.*—A few years ago, in my discussion of Dr. Hollomon's paper, "A Mechanical Equation of State," I referred to some experiments with aluminum, described in an earlier paper.[22] These experiments showed that after aluminum has been plastically deformed at liquid air temperature, there is considerable relaxation when the metal is returned to room temperature. The results thus implied that an equation of mechanical state is not applicable to aluminum between room temperature and −188°C. At that time, I thought it possible that an equation of mechanical state might apply, throughout this temperature range, to metals with higher melting point, such as steel and nickel. However, in a paper published in August, 1947[23] we presented evidence that the flow stress for steels in liquid air is higher when the plastic deformation has been entirely at that temperature, than it is when part of the same total plastic deformation has been at room temperature. This evidence is found in Fig 6, 8 and 11 of that paper. Moreover, in Fig 2 of a paper published in Jan., 1948, we presented similar evidence concerning the flow stress of ingot iron.[24]

During the last four months, I have studied the evidence in a series of our papers. The results have been assembled in 23 full page illustrations, and are discussed in a paper to be offered for publication by AIME. The results, like those presented in the paper by Dorn, Goldberg, and Tietz, show clearly that the rate of real work hardening increases with decrease in the temperature of flow. The results indicated that the conception of an equation of mechanical state is invalid.

---

* Presented as discussion at the Seminar on Cold Working, ASM, Philadelphia, 1948.
† Ibid.

* National Bureau of Standards.
† Case Institute of Technology.

* National Bureau of Standards.
[22] References are at end of discussion.

J. E. DORN, A. GOLDBERG and T. E. TIETZ (authors' reply)—The authors wish to take this opportunity to thank Mr. I. Kramer of the Office of Naval Research for his competent oral presentation of this paper at the AIME meeting at Philadelphia, Oct. 26, 1948. They also express their appreciation to those contributing discussions to this paper, and are gratified to hear that the results of other workers are in substantial agreement with those given here.

The results reported in the text of this paper prove that the concept of equivalent strains is fundamentally untenable. The authors agree with Dr. Hollomon, however, that properly determined equivalent strains may possibly be of considerable practical usefulness for correlations of stress strain data obtained under various strain-rate and temperature histories. On the other hand it must be recognized that such correlations become acceptable only for small strains and that the concept of equivalent strains will not assist in arriving at a better understanding of work hardening.

The comments of G. W. Geil, N. L. Carwile and also Dr. D. J. McAdam on recovery of aluminum at room temperature after straining at liquid nitrogen temperature are appreciated by the authors. Recent tests on pure aluminum (99.987 pct) have shown that the rate of recovery at 90°F is approximately twice as great after straining at liquid nitrogen temperature as after straining at room temperature when the amounts of prestrain at liquid nitrogen and room temperature are so selected as to result in the same value of initial flow stress were the tests immediately continued at liquid nitrogen temperature.

In the present paper the authors reported a test in which a specimen of high purity aluminum was prestrained at room temperature and then tested at liquid nitrogen temperature. From the data given in Fig 14, it is evident that the flow stress for pure aluminum tested at liquid nitrogen temperature after 0.15 prestrain at room temperature falls about 9,000 psi below that obtained when the specimen is strained the same amount at liquid nitrogen temperature. The amount of recovery which occurred after prestraining at room temperature before continuing the test at liquid nitrogen has been determined from a series of recovery tests to be not more than 400 psi, accounting for only a small part of the difference noted. Although recovery of high purity aluminum does occur at atmospheric temperature, such recovery cannot account for the observed failure of the so-called "mechanical equation of state."

Any possible recovery of 2S-0 at atmospheric temperature after prestraining at liquid nitrogen temperature would have decreased the observed difference between the flow stresses of Fig 12 at room temperature. In the absence of recovery, therefore, somewhat greater effects of the past strain history would have been obtained than those recorded in Fig 12. The observed effects of past strain history on the flow stress described in this paper therefore are only slightly affected by recovery. The conclusion that the postulated "mechanical equation of state" is incorrect appears to be inescapable.

References

22. D. J. McAdam Jr., G. W. Geil, and R. W. Mebs: Effects of Combined Stresses and Low Temperatures on the Mechanical Properties of some Nonferrous Metals. *Trans.* Am. Soc. Metals, (1946) 37, 497–537.
23. D. J. McAdam Jr., G. W. Geil, and R. W. Mebs: Influence of Plastic Deformation, Combined Stresses, and Low Temperatures on the Breaking Stress of Ferritic Steels. *Trans.* AIME (1947) 172, 323: *Metals Tech.* TP 2220, Aug. 1947.
24. D. J. McAdam Jr., G. W. Geil, D. H. Woodard, and W. D. Jenkins: Influence of Strain Aging on the Fracture Stress of Low-carbon Steel. AIME *Metals Tech.* TP 2318, Jan. 1948.

The Room and Elevated Temperature Properties of Some Sand Cast Magnesium-base Alloys Containing Zinc

By T. E. LEONTIS, Junior Member AIME

DISCUSSION

(*P. R. Kosting and W. A. Johnson presiding*)

JAY R. BURNS*—The presentation of this paper represents another significant advance in the field of magnesium metallurgy. The higher creep strength of the magnesium-zinc base alloys should place these materials in many applications which stand waiting.

Dr. Leontis has shown a comprehensive picture of the relative merits of many conceivable magnesium-zinc base alloy compositions. Work in the Materials Laboratory has substantiated the good properties obtained from such alloys at room temperature although no elevated temperature testing has been done. One factor which has been evident in the work, however, is the tendency of these alloys to produce a dirty casting. Apart from the microporosity, small oxide skins appear prevalent in cast test bars. The proper use of screens and steel wool obviated the difficulty in the bars but some difficulties might be expected in larger castings. The presence of zirconium was found to emphasize the difficulties of obtaining sound and clean bars whereas Mischmetal reduced the trouble.

Certain other investigators have reported yield strengths of about 28,000 psi and corresponding elongations of 6–8 pct for magnesium-zinc-zirconium alloys. The Materials Laboratory has also obtained such figures from these alloys but the number of tests has not been very large. I should like to ask the author if he believes the figures 24,000 psi and 3 pct for yield strength and elongation are more in line with the typical figures for the alloy.

T. E. LEONTIS (author's reply)—The comments of Mr. Burns are most welcome.

In regard to the tendency of magnesium-zinc alloys to produce dirty castings, I cannot say that we have had the same experience. The binary alloys and all the ternary alloys with which we have worked, except those containing zirconium, do not require any more special treatment than either of the two commercial magnesium sand-casting alloys, AZ63 and AZ92, to produce clean castings. It is true that zirconium does introduce more flux, oxide skins, and nonmetallic inclusions, but here again, either by careful flux treatment and settling or by proper chlorination, clean

---

* Materials Laboratory, Air Materiel Command.

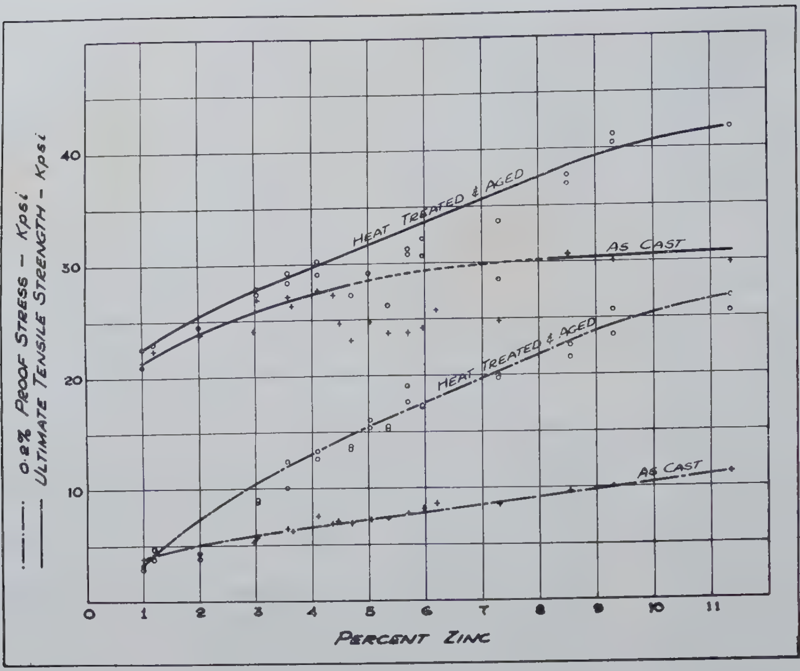


FIG 26—Tensile properties at room temperature for binary Mg-Zn alloys.

castings can be made. We have not experienced any more difficulty in making clean castings of production size than in making clean test bars.

Certain differences also exist between our observations on the microporosity of Mg-Zn alloys and those of Mr. Burns. Binary magnesium-zinc alloys have a high porosity tendency. This tendency is most markedly reduced by the addition of Mischmetal as Mr. Burns states, but we have observed that the addition of zirconium or any of the other elements investigated, with the exception of manganese, also reduces the microporosity to the point where relatively sound castings can be produced. The soundness of these castings compares favorably with that of AZ63 alloy. Alloys containing 0.7 to 2.0 pct manganese, on the other hand, produce very porous castings.

In reply to Mr. Burns' question about the mechanical properties, I do not believe we are in a position as yet to say what the typical properties of Mg-Zn-Zr alloys are. I am assuming that the properties which Mr. Burns has cited refer to an alloy containing 5 to 6 pct zinc and 0.7 pct zirconium. We have also seen 28,000 to 30,000 psi yield strengths and 6 to 8 pct elongations in certain batches of this composition. I used the somewhat lower figures in this paper because those were the room temperature properties of the batch on which I also had elevated temperature data. I consider that these properties are related directly to the grain size. Whether one obtains 28,000 psi, and even higher, or 24,000 psi, and somewhat lower, depends on whether the grain size is 0.002 to 0.004 in. or 0.007 to 0.014 in., respectively. The control of the grain size of these alloys, particularly in production size castings having sections of varying thickness, is a problem that requires considerable further investigation.

J. W. MEIER*—The paper by Dr. Leontis covers a very wide field and represents a considerable amount of experimental work. The author is to be congratulated for his effort and the presentation of the results.

Experimental work on higher strength magnesium alloys at the Canadian Bureau of Mines included also many alloys based on the Mg-Zn system. Since most of our work is still not published, I would like to mention only some of the results which differ from the data presented by Dr. Leontis.

Our results of tensile properties at room temperature for the binary Mg-Zn alloys, both in the "as cast" and the "HTA" conditions, show (Fig 26) an increase of UTS for the zinc range of 8–11 pct to over 30,000 psi and over 40,000 psi respectively, with elongations of 7–8 pct.

These results are based on alloys prepared from high-purity magnesium and electrolytic zinc, without any manganese additions.

The mechanical properties at room temperature for Mg-Zn-Zr alloys, reported by the author, would seem to be somewhat lower than average results achieved in our investigations. As already reported* "as cast" ZK61 alloy test bars showed average UTS of 38–40,000 psi, 0.2 pct PS of 20–22,000 psi and elongations over 10 pct. Grain sizes of the Mg-Zn-Zr alloys, shown in Table 2, reveal that the "effective" zirconium content in alloys 18–24 was rather low. Our investigation showed that the best mechanical properties in these alloys can be achieved only if the zirconium content is well above 0.7 pct and the average grain size is in the range of 0.001–0.002 in., in any case below 0.003 in.

Special heat treatment cycles were established for these alloys which increased considerably the mechanical properties without any effect on the grain size. Maximum results on fully heat treated ZK61 alloy castings of 48,000 psi UTS, of 35,000 psi 0.2 pct PS, and of 7–12 pct elongation were achieved. Considering the strength-to-weight ratio, these alloys have tensile properties superior to any aluminum casting alloys.

Another interesting characteristic of the Mg-Zn-Zr alloys is room temperature aging. "As cast" test bars, stored for some time at room temperature, show considerable increase of proof stress and hardness, as well as some increase of the UTS. Although a more systematic study of this problem is still under way, it may be said that increases of UTS up to 3,000

---

* Canadian Bureau of Mines, Ottawa, Canada.

* J. W. Meier and H. Livingstone: A Study of Magnesium-Zirconium Sand Casting Alloys. Paper presented at the General Meeting, Canadian Inst. of Min. and Met., Vancouver, B.C., Apr. 7, 1948.

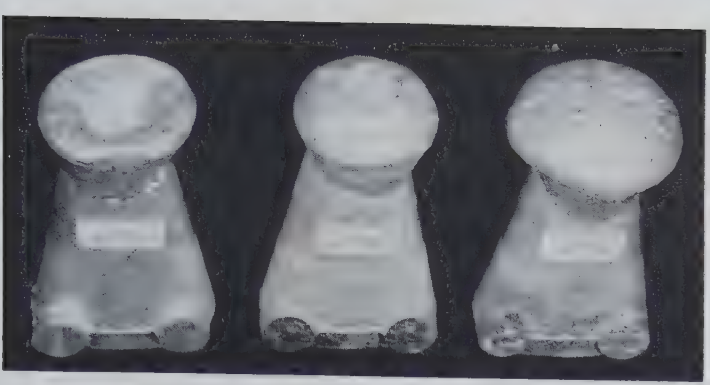


FIG 27—Risers cut from test bar castings—Binary Mg-Zn alloys. Each melt treated with equal amounts of hydrogen compound.

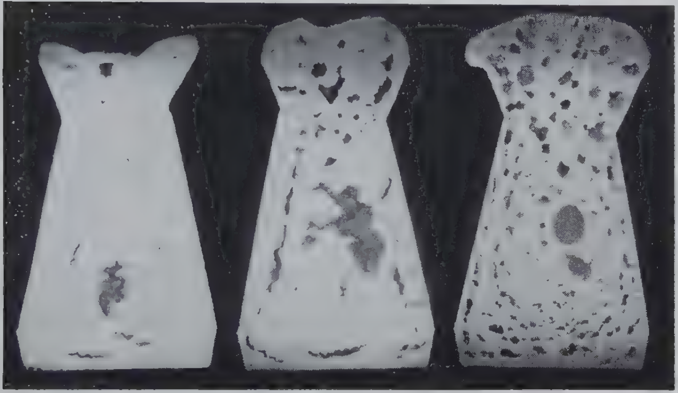


FIG 28—Same as Fig 27, risers sectioned to reveal internal porosity.

psi, and of 0.2 pct PS up to 6,000 psi, have been found.

The effect of calcium additions to Mg-Zn-Zr in the range of 4–6 pct Zn was found to be detrimental to the tensile properties causing a considerable decrease of both the UTS and the elongation.

H. S. SPENCE*—This paper is a valuable contribution to the growing literature of magnesium and its alloys and the author is to be congratulated on both the work accomplished and its presentation.

There is one point which has been mentioned very briefly but which has important practical significance. This concerns the sensitivity to porosity of the binary magnesium-zinc alloys. Mr. Leontis states, in part, "Binary alloys containing more than 2 to 3 pct zinc . . . have a high porosity tendency. . . . All other binary and polynary alloys investigated were essentially free of porosity." Is it correct to infer that the binary magnesium-zinc alloys below 2–3 pct zinc were found to be essentially free of porosity?

It is stated that all alloy melts were prepared according to the 'crucible method' described by Nelson. Was any degassing attempted in the high zinc binary alloys using chlorine or any other agent?

Personal experience in alloying experiments confirms the findings of greater susceptibility to porosity in the higher zinc content binary alloys. This is certainly true with reference to hydrogen and the sensitivity in the range 2 to 6 pct zinc appears to be a function of zinc content. Fig 27 and 28 show this effect quite clearly; these illustrate risers cut from test bar castings from three melts containing respectively 1.89, 3.98 and 6.02 pct zinc. To each melt was added an identical quantity of hydrogen in the form of a metal hydride. The effect of zinc content in repressing the solid solubility of hydrogen is evident. Flushing similarly prepared melts with chlorine did not prove satisfactory in eliminating the hydrogen, even when using a quantity and treatment time in excess of that consistent with practical considerations.

Returning to Dr. Leontis' paper, it is interesting to note that his work does not confirm earlier published data (J. A. Gann,[6] National Physics Laboratory,[6] P. Spitaler[7]) which indicate a fairly sharp maximum UTS of as-cast binary Mg-Zn alloys somewhere in the range between 4 and 6 pct zinc. While the earlier workers were not in agreement as to maximum UTS or its corresponding per cent zinc, all three curves indicated a definite maximum as opposed to the findings now reported which show a constant UTS over the range from 2 to 7 pct zinc. Does the answer to this difference lie in the number of determinations made in the present work with the consequent elimination of erratics, or is it in variations in the foundry techniques used by the different investigators? I suggest it may be the latter since, as these alloys respond to heat-treatment, varying cooling rates after casting may produce effects equivalent to different degrees of partial heat-treatment.

M. W. MARTINSON*—This paper has been a valuable contribution to the magnesium industry, particularly to those associated directly with the development of new and better alloys and should serve as a basis for the further study of magnesium base alloys containing zinc, zirconium and rare earths.

---

* Dominion Magnesium Limited, Haley, Ont., Canada.

[6] References are at end of discussion.

* Dominion Magnesium Limited, c/o Bureau of Mines, Ottawa, Canada.

Mr. Leontis states that, "The addition of 3 pct cerium added to Mg-3Zn improves the strength properties at 300 and 400°F, but is detrimental to room temperature strength. The addition of 3 pct zinc to Mg-3Zn to make Mg-6Zn alloy, however, results in still higher properties at elevated temperatures than those of the alloy containing cerium, and at the same time, is beneficial to the strength at room temperature."

Reviewing the creep data presented for these alloys we find that Mg-3Zn-3Ce is considerably superior to Mg-6Zn at both 300 and 400°F at all loads tested.

Early in our investigations carried out at the Canadian Bureau of Mines in Ottawa, we concluded that zinc had the advantage of raising the mechanical properties of magnesium base alloys intended for elevated temperature use up to 300°F, but above this temperature its presence was detrimental, particularly increasing the creep rate of these alloys.

This raises a question. From the engineering point of view, at what point does the author consider tensile properties to be overshadowed by creep properties in the final evaluation of a magnesium base alloy for elevated temperature use?

In the making of the various magnesium base alloys containing cerium and zirconium: 1. Was it necessary to degas the melt in any manner before or after the zirconium chloride fused salt addition, or was the refining effect from the fused salt mixture sufficient? 2. To what degree did a varying iron content in the Mischmetal added affect the final zirconium analysis? 3. Could the author give an idea of the initial alloying efficiency to be expected for both the rare earths and zirconium additions? 4. What recovery of rare earths and zirconium can be expected on remelt, and what methods are used to revivify the rare earths and zirconium losses to restore the initial properties of the alloy?

C. J. P. BALL*—My metallurgical staff and I were greatly interested in the paper by Mr. Leontis and I attach some detailed comments by my staff upon certain points in which they are particularly interested. Since Mr. Leontis touches upon the development and usage of magnesium-zirconium alloys, I think you would be interested to learn the present position in the United Kingdom as regards the magnesium-base casting alloys containing zinc and zirconium with and without Mischmetal.

At our Clifton Junction works we have carried out, over a number of years, a considerable volume of research and development work on these zirconium-containing alloys. Two major problems presented themselves, and both had to be solved at a reasonable cost. The first was the introduction of sufficient zirconium in effective form without serious chloride contamination, the second, the suppression of microporosity, which has always presented a danger with all forms of magnesium alloys, and which can also ruin castings in high strength zinc-zirconium containing alloys unless the correct technique is used.

I am glad to say that special alloying compounds and fluxes have been evolved by my company, which enable the alloying to be carried out simply and at a reasonable cost, with no danger of flux contamination, and which yield a zinc-containing alloy virtually free from microporosity even in the thickest sections. As a result this alloy, Z5Z, (4.5 pct zinc, 0.7 pct zirconium) is being called for in rapidly increasing quantities for many structural castings, particularly in the aircraft industry. The properties of the casting are remarkably uniform, and frequently exceed those of separately cast test bars. Because of the ease with which this alloy can be handled in the foundry and the superior mechanical properties at room temperature, it is expected that this alloy will replace the greater part of the magnesium-aluminum-zinc alloys at present used for structural purposes.

The advances made recently in the magnesium-base alloys containing Mischmetal in addition to zirconium and zinc have been most striking. Because of their resistance to creep at temperatures around 200°C, their main applications so far have been elevated temperature work. The ZRE1 alloy (2.7 pct zinc, 2.7 pct Mischmetal and 0.7 pct zirconium) has made particular progress. Its properties at room temperature are—

Yield —14,000 psi
Tensile strength—23,000 psi
Elongation —5 pct

Around 200°C its resistance to creep is practically identical with that of RR.50, the well-known general-purpose aluminum-base alloy widely used in the United Kingdom for service at elevated temperatures. Incidentally, the tensile strength of 15,000 psi given in the recent American Specification AMS.4428 would seem to compare most unfavorably with the corresponding figure for ZRE1, particularly since the small margin of 2,000 lb only between tensile and yield indicates the inherently brittle nature of this alloy at room temperature in the cast state.

Practical experience in the fabrication of the largest and most complicated castings, even as high as 900 lb net weight, has proved conclusively that the ZRE1 alloy containing not less than 2 pct Mischmetal is inherently free from any tendency to microporosity.

This concludes my short resume of the development and present position of the magnesium-zirconium-zinc and magnesium-zirconium-zinc-Mischmetal alloys in this country.

Writing as a member of the AIME, it is my hope that this survey will be found of interest to Mr. Leontis and to those who have taken part in the discussion.

As we understand that considerable work on magnesium-zirconium-zinc and magnesium-zirconium-zinc-Mischmetal alloys is also being carried on in the U.S., perhaps some of the members taking part in the discussion would be able to answer the following questions: 1. Are magnesium-zirconium containing castings now being produced readily free from flux inclusions? 2. Can large castings be made readily in magnesium-zinc-zirconium free from microporosity? 3. Are alloys of magnesium-zirconium-zinc-Mischmetal readily available to the trade?

A. C. JESSUP* and J. B. WILSON*—Mr. Leontis has presented extensive data on magnesium alloys, and since his paper summarizes the properties for many alloys for which published values are not otherwise available, it provides most interesting reading for all workers in the magnesium industry. The article is, however, a little difficult to follow, and the comments and figures occasionally appear somewhat contradictory, nor do we in all cases agree with the inferences which have been drawn.

For instance, Fig 29 shows the effect of adding calcium in amounts up to 1 pct to a Mg-Zn-Zr alloy; in our opinion this shows calcium to be positively injurious. We would agree that the addition of cerium produces a similar reduction in room temperature properties, but for high temperature service the advantage of enhanced creep resistance more than outweighs the disadvantage of lower room temperature properties.

On p. 19 of his paper, the author claims that the Mg-6Zn-0.5Mn alloy is comparable for strength and ductility at all temperatures with the Mg-6Zn-1Zr alloy, and quotes Fig 10. We are not able to substantiate this sweeping statement. While we do not recommend the use of Mg-Zn-Zr alloys at temperatures above 300°F, and favor a lower Zn content, our experience based upon hundreds of tests and many tons of production, has satisfied us that the room temperature properties of a Mg-4.5Zn-0.7Zr alloy when prepared with the proper alloying reagents are certainly higher than the author quoted for Mg-6Zn-0.5Mn at room temperature. In Fig 10, the room temperature properties of the manganese containing alloy are given as:

Yield —24,000 psi
Tensile strength—38,000 psi
Elongation —5 pct

---

* Magnesium Elektron Limited, London, England.

* Magnesium Elektron Limited, Clifton Junction, Manchester, England.

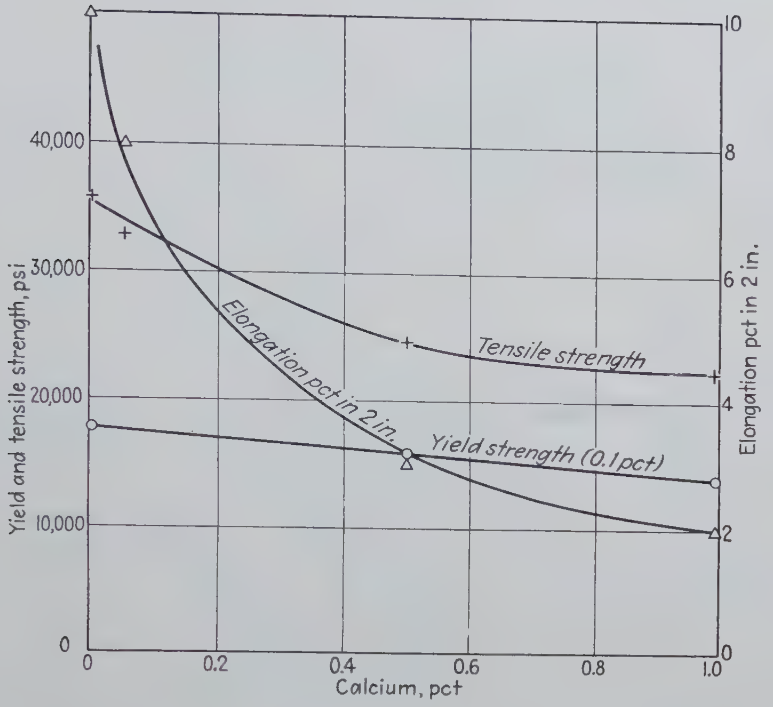


FIG 29—Effect of adding calcium in amounts up to one per cent to Mg-Zn-Zr alloy.

Comparable properties for a Mg-4.5Zn-0.7Zr alloy when tested as an unmachined "American" type test bar, after an annealing treatment of 10 hr at 350°F, are:

Yield —27,000 psi
Tensile strength—40,000 psi
Elongation —5 pct

Moreover, pronounced freedom from microporosity and grain coarsening on heat treatment, and small variation in mechanical properties between test bars and specimens cut from the casting, make this alloy particularly attractive to the foundryman. We have not had much experience with the 6 pct zinc-0.5 pct manganese alloy, but such as we have had has shown it to be even more microporous than AZ92, which in turn suffers much more from microporosity than the 4.5 pct zinc-0.7 pct zirconium alloy.

The improvements offered by the Mg-Zn and similar alloys over the AZ92 and AZ63 alloys are noted with interest. We feel, however, that the all-round properties of these alloys do not offer to magnesium founders the great advances and improvements in properties and casting techniques which they can obtain from the Mg-Zr-Ce alloys developed by Murphy and Payne or from the alloys of Mg-Zn-Zr-Ce, designated ZRE1 and ZRE2, recently developed in the United Kingdom by Magnesium Elektron Ltd. These latter contain nominally Mg-2.7Zn-0.7Zr-2.7Ce and Mg-4.5Zn-0.7Zr-2.0Ce, respectively. Mr. Leontis has tested two closely similar compositions (his alloys 28 and 30) and finds them among his most promising alloys at 300°F. It was surprising to find therefore, that he did not investigate this type of alloy in more detail at 400°F, particularly in view of their extremely attractive combination of properties over the 60–400°F range. In Fig 30, we have plotted several of the author's results from Tables 6 and 7 at 300 and 400°F and have added a few of our own obtained at 400°F. In this diagram we consider *creep rates* only, since, in our opinion, the use of an arbitrary "creep limit" is likely to prove misleading.

Our results include the ZRE1 and ZRE2 alloys and also the Mg-3Ce-0.7Zr composition (MCZ) as developed by Murphy and Payne. We find that there is little to choose between the MCZ and ZRE1 alloys over the 60–400°F range of temperatures. The zinc containing alloys are, however, distinctly easier to handle in the foundry. The superiority of the cerium bearing alloys at 400°F requires no comment.

The author states on p. 29 and 31 of the paper that zinc lowers the creep resistance of Mg-Ce alloys. This may well be true in the particular case of zinc additions to the binary Mg-Ce alloy. However, all our experience has shown both in tests and in practical foundry production that if the full amount of effective zirconium is present, considerable amounts of zinc may be tolerated without adverse effect upon creep resistance. This is demonstrated by Mr. Leontis' own experience with compositions 28 and 30 and our experience with the alloys ZRE1 and ZRE2. This may be due to the suppression of the formation of zinc phase when zirconium is present, which would in turn prevent the "dilution" of the cerium compound by zinc compounds at the grain boundary, and would enable it to act as if the zinc were absent. If Mr. Leontis did not find the Mg-Ce-Zn-Zr alloy quite as satisfactory as the Mg-Ce binary for creep, then judging from the results he quotes for other zirconium containing alloys, it would seem to be probable that his results are unsatisfactory because he has not been able to introduce into his alloy the necessary amount of effective zirconium to counteract the zinc.

In conclusion, we should like to record that these differences of experience and interpretation apply only to a small part of the paper, the majority of which is entirely in accord with our own work and experience.

T. E. LEONTIS (author's reply)—It is

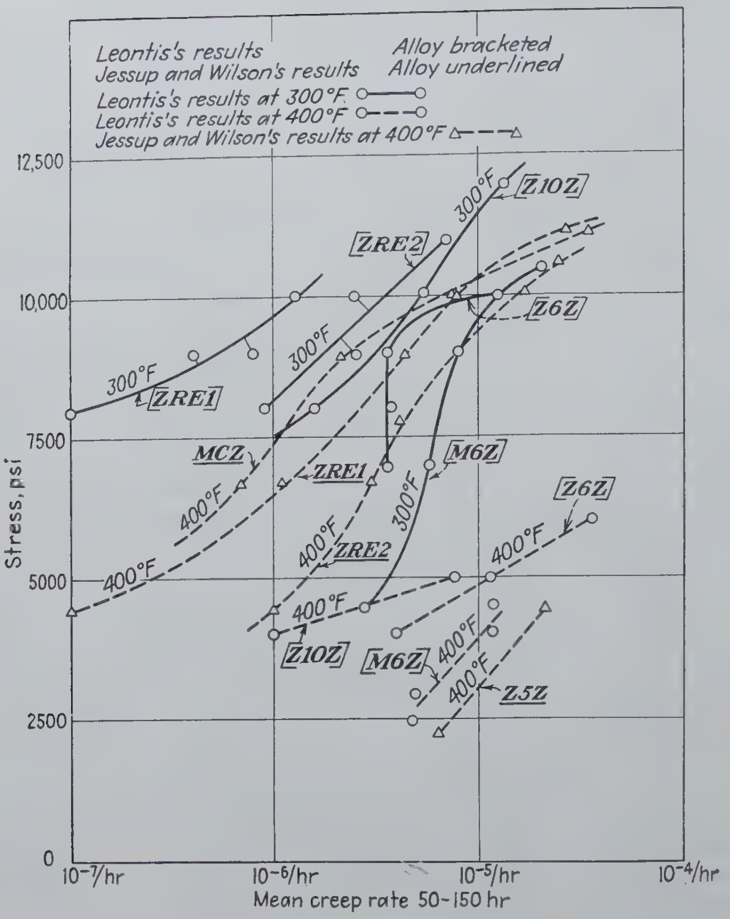


FIG 30—Plot of results from Tables 6 and 7 at 300° and 400°F, and others at 400°F.

gratifying to see that this paper has aroused so many interesting discussions from our friends in Canada and Great Britain. In the following, I shall attempt to present briefly my views on the many questions raised by the various contributors.

Mr. Meier and Mr. Spence have mentioned the properties of binary Mg-Zn alloys. It is interesting to note that the tensile yield strength of their alloys both in the as-cast and in the heat treated and aged condition, as shown in Fig 26 of Mr. Meier's discussion, are in reasonably good agreement with the values given in Fig 6, particularly if the curve for the heat treated and aged condition is drawn through all the points as is done in Fig 6. In regard to the tensile strength, if one looks at the experimental points rather than the curves drawn in Fig 26, again the agreement is good between the two sets of data up to 7.5 pct zinc. The absence of a sharp maximum in this property between 4 and 6 pct zinc that was observed by earlier investigators, as mentioned by Mr. Spence, is noted also in Fig 26. The markedly higher values depicted by Mr. Meier's results in alloys containing 8 to 11 pct zinc must be associated with differences in foundry practice. Perhaps the conditions of preparing the alloys were such as to result in a finer grain size and, what is more important in alloys containing a large amount of zinc, in a finer distribution of the intermetallic compound. Other factors that also may affect the tensile strength of Mg-Zn alloys have been cited by Mr. Spence. In reply to the question about the porosity of binary Mg-Zn alloys, my observations of freedom from porosity in alloys containing up to 2–3 pct zinc were made on cast test bars. The evidence that Mr. Spence has presented indicates that difficulty can be expected in larger casting even in the lower zinc contents. Attempts made to minimize porosity in Mg-6Zn alloy by chlorination have not been very successful.

In line with the observations of Mr. Burns, Mr. Meier has reported some still more spectacular properties for cast ZK60 alloy. Their high strength and ductility values are associated undoubtedly with the very fine grain size that Mr. Meier was able to achieve in his alloys. As he has noted, in many of the alloys in the paper the zirconium content mentioned is below 0.7 pct, an amount which he feels must be exceeded in order to attain the finest possible grain size. However, alloys prepared recently in which the zirconium content exceeded his minimum have not confirmed his findings either with respect to grain size or properties.

Turning to the comments of Major Ball concerning Mg-Zn-Zr alloys, I was very interested to hear of his successful application of these alloys in the foundry. We have been successful in producing certain castings in ZK60 alloy free from flux, porosity, and cracking. The flux problem can be eliminated by proper treatment as discussed above. However, difficulties are sometimes encountered with porosity and cracking when more intricate castings are made from this alloy. The ingredients for making up these alloys are freely available in the United States on the open market.

The question of room temperature strength vs. elevated temperature strength and creep resistance raised by

Table 10 . . . Comparison of Alloys

Temp. °F	Mg + 6Ce			Mg + 3Ce			Mg + 3.1Zn + 2.7Ce + 0.45Zr		
	Pct E	TYS	TS	Pct E	TYS	TS	Pct E	TYS	TS
		1000 psi			1000 psi			1000 psi	
Room	1.0	17.0	20.4	1.5	12.4	20.5	3.8	13.4	22.3
300	2.0	14.9	19.3	7.0	12.9	17.9	14.4	11.9	21.7
400	2.0	14.7	18.6	4.7	11.9	16.8	19.8	11.6	18.4
500	4.0	13.4	17.9	11.7	11.3	17.9	44.0	9.7	14.4
600	12.5	9.8	14.3	24.2	8.8	13.5	66.5	6.7	9.1

The data are self-explanatory.

Mr. Martinson is one which is frequently asked but it is not answered very often, at least not to the satisfaction of designers and engineers. Certainly, if a part is to be exposed to sustained loads of predetermined value at elevated temperatures, the resistance to creep at those temperatures would necessarily be the controlling factor in design. However, it is always necessary to make provision for overloading which may occur during operation or during assembling of the part. For this reason, it will be found that the highest possible static properties are desired both at room temperature and at elevated temperatures in order to reduce failures to a minimum. Thus, in the selection of an alloy for elevated temperature service, a compromise may have to be made between static properties and creep resistance. The answers to questions of Mr. Martinson concerning foundry practice are as follows: 1. This has been answered adequately in reply to Mr. Burns' discussion. 2. We have no data from a systematic investigation of this effect. There unquestionably is a lowering of zirconium content resulting from the introduction of iron when Mischmetal is added to the melt and it certainly would be expected that the degree of lowering will be dependent on the amount of iron added. 3. The initial alloying efficiencies of rare earths and zirconium are:

	Pct
Rare earths	80–90
Zirconium	20–40

4. The losses of rare earths and zirconium in absolute amount upon remelting are approximately:

	Pct
Rare earths	0.5
Zirconium	0.1–0.2

The revivification of rare earths and zirconium is accomplished by adding the amount lost after remelting. The additions are made on the basis of the alloying efficiencies given above.

The discussion of Messrs. Jessup and Wilson contains many interesting points, some of rather controversial nature. I was interested to see their results on the effect of calcium on the properties of Mg-4.5Zn-0.7Zr alloy. It will be noted in the paper that the claim for an enhancement of the properties by the addition of 0.3 pct calcium was made only for the composition Mg-1Zn-0.7Zr. Fig 9 of the paper shows that the same addition of calcium to Mg-3.5Zn-0.7Zr results, if anything, in a slight decrease in room temperature tensile strength. It is apparent from these data and those of Messrs. Jessup and Wilson that the effect of a small addition of calcium to these alloys depends upon the zinc content; at low zinc contents there is an increase in strength which appears to go through a maximum between 0 and 3 pct zinc, and at zinc contents above 3–3.5 pct zinc there is a definite decrease in strength resulting from the calcium addition.

The properties of Mg-6Zn-0.7Zr (ZK60) and Mg-6Zn-0.5Mn (ZM60) are more readily compared in Table 9. With the data available at the time of the writing of the paper, I cannot see any great differences between the properties of the two alloys with the exception of the somewhat lower ductility of ZM60 at 300 and 400°F. It is true that higher strengths can be attained in ZK60; in fact, considerably higher properties than those cited by Messrs. Jessup and Wilson have been shown by Mr. Meier. As for the foundry characteristics of the two alloys, the statements made in the paper were based on observations from the casting of test bars and a few relatively simple castings. Since the writing of the paper, more extensive tests on the foundry characteristics both of ZK60 and ZM60 have shown that the latter alloy has a much greater porosity tendency in larger castings than the zirconium-containing alloy.

Both Major Ball and Messrs. Jessup and Wilson have indicated that for elevated temperature service they would favor an alloy of the Mg-Zn-Ce-Zr type. It has been the thesis of this paper that Mg-Zn-Zr alloys offer a combination of room temperature properties equivalent to those of the present Mg-Al-Zn commercial alloys and significantly higher creep resistance at elevated temperatures than the commercial alloys. Mg-Al-Zn alloys have been used widely in parts exposed to temperatures up to 300°F. With the higher resistance to creep of the Mg-Zn-Zr alloys, there is every reason to believe that these alloys would be satisfactory in many parts operating at temperatures as high as 400°F.

For applications at temperatures above 400°F, we in the United States have been recommending Mg-Ce alloys. Major Ball's comparison of the laboratory properties of ZRE1 alloy with those of E6 (Mg + 6 pct Ce) alloy as specified in AMS4428 is not proper. Specified minimum properties are always set lower than typical values. A more realistic comparison can be made by using laboratory data for both alloys. This is done in Table 10 where the properties of an alloy having an analysis similar to that of Major Ball's alloy and those of two Mg-Ce alloys are presented over a wide temperature range. All these alloys are in the heat treated and aged condition.

It is the contention of Messrs. Jessup and Wilson that the elevated temperature creep resistance of Mg-Zn-Zr-Ce alloys is equivalent to that of the Mg-Ce alloys. Since the writing of the paper a few alloys of the former type have been tested in creep at 400°F. The results of these tests together with the corresponding properties of two Mg-Ce alloys are tabulated below. All the alloys were tested in the heat treated and aged condition. In order to present a complete picture, the creep limits based on three levels of extension are given in Table 11: (1) 0.1 pct creep extension, (2) 0.2 pct total extension, and (3) 0.5 pct total extension. I prefer the use of a creep limit based on a specified amount of extension to the use of an actual creep rate for several reasons. In the first place, a creep limit can be used directly in design calculations. Secondly, during creep tests of short duration, the creep rate very often does not attain a constant value. Thus, such creep rates are unreliable for predicting behavior in practice. It can be seen from the above summary that the best creep resistance is offered by Mg + 6 pct Ce alloy. This fact, coupled with the superior strength of this alloy at elevated temperatures, renders the material more attractive than the Mg-Zn-Ce-Zr alloys for applications up to 600°F. On the basis of the above data, particularly the stress values for 0.2 and 0.5 pct total extension, it appears that the alloy ZRE1 approaches the creep resistance of Mg + 3Ce. Inasmuch as Messrs. Jessup and Wilson have presented only creep rates in their discussion, it would be interesting to compare their figures with those obtained in the present tests. The values for the Jessup and Wilson alloys given in Table 12 were taken from their Fig 30. There are apparently wide discrepancies between the two sets of data. If creep rate were used exclusively for evaluating the creep characteristics of the alloys, the present author's results indicate a very significant superiority of the Mg + 3Ce alloy over the zirconium-containing alloys.

Table 11 . . . Creep Limits Based on Three Levels of Extension

Alloy	Creep Limit — 1000 psi (400°F)		
	0.1 Pct Creep in 100 hr	0.2 Pct Total Extension in 100 hr	0.5 Pct Total Extension in 100 hr
Mg + 6.33Ce	10.7	9.6	11.8
Mg + 3.23Ce	10.0	8.3	9.9
Mg + 3Zn + 2.8Ce	7.8	5.7	8.4
Mg + 3.1Zn + 2.7Ce + 0.45Zr (ZRE1)	7.7	7.0	9.1
Mg + 4.4Zn + 1.7Ce + 0.66Zr (ZRE2)	5.8	5.8	7.4
Mg + 4.4Zn + 2.8Ce + 0.57Zr	6.4	6.2	8.0

Table 12 . . . Values of Alloys

Creep Rate per hr (400°F)					
Mg + 3.1Zn + 2.7Ce + 0.45Zr ZRE1 Alloy			Mg + 4.4Zn + 1.7Ce + 0.66Zr ZRE2 Alloy		
Stress psi	Present Data	Jessup and Wilson	Stress psi	Present Data	Jessup and Wilson
7000	3.9×10^{-6}	1.4×10^{-6}	6000	6.8×10^{-6}	2.4×10^{-6}
9000	16.9×10^{-6}	4.6×10^{-6}	7000	18.8×10^{-6}	3.3×10^{-6}
11000	302.0×10^{-6}	20.0×10^{-6}	9000	∞	4.7×10^{-6}
Mg + 3.23Ce Alloy			Mg + 3.1Zn + 2.7Ce Alloy		
8000	1.0×10^{-6}		5000	0.42×10^{-6}	
9000	8.7×10^{-6}		6000	1.2×10^{-6}	
10000	9.0×10^{-6}		8000	4.4×10^{-6}	
11000	68.0×10^{-6}				

There is one factor that has not been considered as yet. Whereas all the results that I have presented were obtained on alloys in the heat treated and aged condition, there is no mention of any heat treatment made by Messrs. Jessup and Wilson. Presumably, all their tests were performed on material in the as-cast or in the as-cast plus aged condition. At the present time, I have creep data available on Mg + 4.4Zn + 3.7Ce + 0.55Zr at 300°F in both conditions. These are presented in Table 13 together with the corresponding information for Mg + 3.23Ce. On the basis of creep limits there is little to choose between the two conditions of heat treatment, but the creep rates show the ACA condition to render the alloys somewhat more resistant to creep. The Mg-Ce alloy is again superior to the zinc-containing alloy both on the basis of creep limits and of creep rates in spite of the somewhat lower cerium content. This information indicates that the lower creep rates reported by Messrs. Jessup and Wilson may be attributed to their testing the alloys in the ACA condition. However, this emphasizes the hazards of using only the creep rate in evaluating alloys for elevated temperature service. Inasmuch as the total extension obtained is the same in both cases, the lower creep rate of the alloys in the ACA condition means that more of the extension is obtained during loading and less during the creep test.

It has been proposed that perhaps the lower resistance to creep of the Mg-Zn-Ce-Zr alloys used in the present paper may be due to our inability to introduce a sufficient amount of zirconium in our alloys. The zirconium content of most of my alloys is admittedly lower than that claimed by Messrs. Jessup and Wilson as being necessary for attaining the highest possible creep resistance. However, reference to Table 1 in the paper shows that alloy No. 28 has the lowest zirconium content of the cerium-containing alloys and at the same time has been shown to have the highest creep resistance. In fact, the creep rates of the alloy with no zirconium given above are lower than those of either of the two zirconium-containing alloys. The data show that high creep resistance is associated with a high Ce/Zn ratio emphasizing again the predominant position of cerium in magnesium alloys for elevated temperature creep resistance. As for the necessity of the presence of zirconium to suppress the formation of the zinc intermetallic compound, it has been my observation that even small amounts of cerium alone added to alloys containing up to 6 pct zinc will produce this effect.

In conclusion, only time and further experience can settle current differences of opinion on the preferred magnesium alloys for either room or elevated temperature service. Relative costs based on the alloying elements used, and the foundry efficiencies attained, will certainly be a factor. At elevated temperatures, the actual temperature concerned, and the appropriate creep strengths in the best condition of heat treatment, will affect the decision. An urge for simplicity in foundry operations may result in a compromise of a few alloys of good average properties, rather than more alloys each with its specialized field of use.

I would like to thank all the contributors to the discussion of this paper for their stimulating comments and criticism. I shall be looking forward to seeing the results of their investigations in formal papers.

References

6. J. L. Houghton and W. F. Prytherch: Magnesium and Its Alloys. H. M. Stationery Office, London (1937) p. 35, Fig 10. (Original from *Ind. and Eng. Chem.* (1927) 19, 1193.) National Physical Lab. Ibid, p. 38, Fig 15.
7. A. Beck: The Technology of Magnesium and Its Alloys. F. A. Hughes and Co., Ltd., London (1940), p. 133, Fig 168.

Table 13 . . . Creep Data on Mg + 4.4Zn + 3.7Ce + 0.55Zr at 300°F

	Creep Limit—1000 psi (300°F)		
	0.1 Pct Creep in 100 hr	0.2 Pct Total Extension in 100 hr	0.5 Pct Total Extension in 100 hr
Heat Treated + Aged (HTA)	11.2	8.1	11.3
As-cast + Aged (24 hr at 350°F (ACA)	11.5	7.8	11.8
Mg + 3.23Ce (HTA)	13.0	10.0	12.5

Stress psi	Creep Rate per hr (300°F)		
	Mg + 4.4Zn + 3.7Ce + 0.55Zr		Mg + 3.23Ce
	HTA	ACA	HTA
8000	0.6×10^{-6}	0.2×10^{-6}	
10000	0.8×10^{-6}	0.4×10^{-6}	0
11000		0.4×10^{-6}	0.2×10^{-6}
12000	7.3×10^{-6}	0.6×10^{-6}	0

The Low Temperature Properties of Tin and Tin-lead Alloys

By H. S. KALISH and F. J. DUNKERLEY, Junior Members AIME

DISCUSSION

(P. R. Kosting and W. A. Johnson presiding)

R. I. JAFFEE*—I should like to amplify the work of Kalish and Dunkerley on the basis of some work we did in conjunction with the National Lead Co. about three years ago, which is to be published in *Metal Progress* early in 1949.

In general, we confirmed the statement that 50–50 lead-tin is brittle at low temperatures. However, we investigated the lead-rich range, and found, as might be expected, that the lead plus 2½ pct silver alloy and lead-rich lead-tin alloys had rather good low temperature properties, both in ductility, strength, and in Charpy impact values.

One observation made on alloys between the high tin range investigated by Kalish and Dunkerley and the high-lead range was that alloys with about 15 pct tin did not decrease materially in impact strength or in ductility, as temperature went down to −295°F. This point, it seems to us, should be emphasized, since the presence of tin, at least up to 15 pct, in a lead-tin solder is rather important for commercial applications because of the improvement tin makes in solderability and wetting power.

J. R. LOW, JR.†—The authors have presented a very interesting paper and one of the few describing a systematic investigation of the relationship between microstructure, composition and low-temperature brittleness in a series of alloys.

The only comment I should like to make is that I believe the use of the term "ultimate tensile strength" over the full range of the curve so labelled in Fig 3 of the paper is misleading. As far as engineering design is concerned, this is a valid use of the term; however, in thinking about the mechanical properties of the metal, a distinction should be made between the quantity measured above and below the transition temperature. Above the transition temperature the ultimate tensile strength is a measure of the stress at which necking begins and is therefore related to the rate of strain hardening. Below the transition temperature, the fracture strength of the metal is being measured. For this reason the maximum ultimate tensile strengths referred to throughout the paper are in a sense fictitious maximums.

H. S. KALISH and F. J. DUNKERLEY (authors' reply)—The authors wish to thank Drs. Jaffee and Low for their comments on this paper. From the results reported here, we would predict a lead-silver alloy containing 2½ pct silver would remain ductile at low temperatures and we are glad that Jaffee's work confirms this as well as our other work on the tin-lead alloys. It might further be speculated in agreement with Brick and Seigle[16] that any alloy formed by the combining of any two or more face-centered cubic metals, such as silver and lead, should remain ductile to absolute zero.

The superior ductility of the 85–15 lead-tin alloy is indeed important as emphasized by Dr. Jaffee. Again this would be predicted if the T_B curve of Fig 11 were extrapolated to higher lead contents. Kostenets[17] found that a tin-lead alloy containing 25 pct tin still remained ductile at −253°C, only 20°C above absolute zero.

We agree with Dr. Low that it would have been desirable to separate the so-called ultimate tensile strength above T_B from that measured quantity below T_B. However, since we did not have true stress-strain, yield strength, work hardening moduli and other fundamental data available, we were not able to separate all the variables quantitatively and preferred to present the engineering tensile data in the form most useful to engineers. These necessary fundamental data, however, are being obtained in this laboratory so that a more rigorous analysis of these low temperature properties can ultimately be made.

Dr. Low is quite right that the rate of strain hardening is the most important factor in determining the tensile strength above T_B, while below T_B the fracture strength is the main component. In neither case, however, is the strain hardening or the fracture stress, respectively, the only component. For example, the ultimate tensile strength, Fig 3, could not be labelled fracture strength below T_B because considerable ductility remained to make an appreciable work hardening contribution to the tensile strength feasible. The necessary qualitative rationalizations of the tensile data of this survey report should be clarified and amplified by the forthcoming fundamental data currently being obtained in this laboratory.

References

16. L. Seigle and R. M. Brick: Mechanical Properties of Metals at Low Temperatures; A Survey. *Trans.* Amer. Soc. for Met. (1948), **40,** 813.
17. V. I. Kostenets: Mechanical Properties of Metals under Static Load at Low Temperatures. *Jnl. Tech. Phy.* (U.S.S.R.) (1946), **16, I,** IV, 515.

---

* Battelle Memorial Institute.

† General Electric Co.

Thermal and Electrical Properties of Ductile Titanium

By E. S. GREINER and W. C. ELLIS, Members AIME

DISCUSSION

(P. R. Kosting and W. A. Johnson presiding)

P. DUWEZ*—We would like first to congratulate the authors for a very interesting piece of work on a difficult subject. We have been studying the thermal expansion of pure titanium for some time and have found the coefficient of expansion between room temperature and 800°C to agree closely with the value given by the authors. We are particularly interested, however, in the thermal expansion anomalies accompanying the $\alpha \rightleftarrows \beta$ allotropic transformation.

On specimens machined out of a bar

---

* Jet Propulsion Laboratory, California Institute of Technology.

manufactured by the vapor deposition process (obtained from Foote Minerals) we have observed a contraction on heating and an expansion on cooling, both around 880°C. The change in length varied somewhat from sample to sample, but was about 0.05 pct. This anomaly in the expansion curve was almost negligible on heating and completely absent on cooling when the bar was cold worked (reduction of area about 30 pct) before the test. The same experiments made on sintered titanium (from Bureau of Mines powder) did not reveal any thermal expansion anomalies.

We would like to ask the authors if they have made any measurements of thermal expansion across the α, β critical temperature and we would appreciate very much their comments on the subject.

W. C. ELLIS and E. S. GREINER (authors' reply)—The discussion by Professor Duwez of the volume changes in the 885°C transformation as evidenced by linear expansion measurements is particularly interesting and the additional data are welcomed. The authors made similar measurements on Bureau of Mines titanium (magnesium-reduced) but did not report the results in the paper because of uncertainty as to their significance. In these experiments there was always a volume decrease on heating through the transformation but no detectable discontinuity on cooling. The result was a shortening of the specimen as it was cycled through the transformation. The cause of the shortening was tentatively assigned to a further sintering of the expansion specimen in traversing the transformation. This was possible since the expansion specimen was only approximately 75 pct reduced in area from the sintered compact and could reasonably be expected to contain a distribution of fine voids.

Sintering of the expansion specimens during test in the temperature range below 800°C, for which data were given in the paper, was insignificant, since the slopes of expansivity on heating and cooling in this temperature range were nearly identical. At the transformation temperature, however, because of the spontaneous rearrangement of atoms the facility for sintering appears to be greatly increased.

B. W. GONSER*—The authors did a very fine job, and everything is very much in order for the work that was covered.

My discussion is a little broader than the subject covered by the paper, but it is activated by noting the purity or type of titanium used. This whole subject of titanium purity is becoming very involved because of the difficulty of defining the metal which is used as the base for many fundamental investigations.

The Bureau of Mines powder, of course, is not nearly as pure as that produced by iodide refining. Likewise, even iodide titanium varies quite widely in analysis. As measured by hardness, for example, it may go all the way from a Vickers hardness of 60 to 115. Therefore one should define very closely the base material used to give an indication of what the properties are going to be. This is involved because of the difficulty of determining many constituents. Particularly the nonmetallics, as oxygen, nitrogen, or carbon, are difficult to determine, and yet are very profound in their effects.

I think if more fundamental work is done on titanium it should be done with very well defined iodide titanium, or even purer metal if there is a better method of making it, to get consistent results on a reasonably sound base.

This choice of base material is important because so much is being done with titanium on different bases and results can be confusing. A lot of work is probably going to have to be done over again from time to time with titanium of improved purity.

J. T. MILEK*—Did the authors observe the phenomenon which J. D. Fast and DeBoer, I believe, observed in their experiments with zirconium wire, that is, by impressing a dc current voltage on a zirconium wire one can change the wire from an embrittled wire to a ductile wire? The explanation given by the authors is that the oxide and/or nitride which cause the embrittlement in the wire are transported to one pole—I believe it was the cathode. While the above work was carried out on zirconium, did the authors observe the same phenomenon with the titanium wire?

Communicated: This phenomenon was observed by J. H. DeBoer and J. D. Fast and reported in Rec. Trav. Chim., Vol. 2, pp. 161–167, (1940). DeBoer and Fast reported that as a piece of brittle zirconium containing oxide is placed under a high vacuum and a dc potential of such a magnitude that the wire attains a temperature of approximately 900°C, the oxygen will migrate through the bar and collect at the anode. This migration can be reversed any number of times by changing the polarity of the current. Ductility can be restored through the cathode by such treatment.

While the above phenomenon has been observed in zirconium metal it is not unreasonable to assume that it can occur in titanium metal, or perhaps by a stretch of the imagination, even in titanium alloys. This latter process may find commercial applications.

W. A. JOHNSON†—I find the results given in Fig 3 for the variation of the resistance with temperature rather interesting, since above the transformation temperature the resistivity seems to decrease with rising temperature.

This is a rather uncommon behavior, and I wonder whether possibly the transformation does not occur quite as rapidly as is indicated, and this decrease in resistivity perhaps is simply a gradual transformation with an increasing amount of a lower resistance phase.

W. C. ELLIS—Dr. Gonser brought up the matter of the purity of the metal which was investigated. It was not high purity in the sense of some of our more usual metals, as shown by the analyses in the paper. I agree with Dr. Gonser that it is desirable to carry out fundamental investigations on metals of high purity. However, in view of the increasing commercial importance of titanium and the fact that the methods which produce the high purity titanium are expensive and have produced only small quantities in a lot, we think there is good reason to have data on the properties of magnesium-reduced titanium.

Mr. Milek brought up a very interesting fact. Ionic transport is not only found in impure zirconium, but is of general interest in other alloy systems. We ordinarily think of electrical conduction as electronic in metals, but in some metal systems there is good evidence that there is a transport by one of the ions. Data have been reported on transport of carbon in steels by applying dc voltage.

In answer to Mr. Milek's question, no experiments of this character were made in our work on titanium.

Mr. Johnson discussed the unusual occurrence of a decrease in electrical resistivity with temperature in the high temperature phase of titanium. The authors are also mystified by this, but believe that it is a real effect, and not the result of a slow transformation. On cycling the specimens through the transformation a number of times, the temperature of the transformation as shown by the electrical resistivity data occurred always at the same level.

J. L. WYATT*—In line with this discussion which has been completed, would you mind telling us how long you held your specimens at temperature before taking your measurements above 885°C.

E. S. GREINER—The sample used for the electrical resistance measurements at temperatures above 885°C was maintained at the respective temperatures for 15 to 30 min. Stabilized conditions for the measurement at the respective temperatures were established by the obtaining of essentially constant values of the resistance in two successive determinations.

---

* Battelle Memorial Institute.

* Babcock and Wilcox Co.
† Westinghouse Research Laboratories.

* National Lead Company.

Property Changes during Aging

By A. H. GEISLER Junior Member AIME

DISCUSSION

(*P. R. Kosting and W. A. Johnson presiding*)

P. MELARA[*]—Dr. Geisler, would you say that the physical properties of tensile, yield, and ductility for an aluminum alloy aged at room temperature are comparable to the physical properties obtained by the aging at elevated temperature of the same alloy? I would consider the aging at room temperature to be for an indefinite period of time, for example, ten, twenty, or even fifty years.

A. H. GEISLER (author's reply)—I think that the properties which could ultimately be obtained by aging at room temperature can be predicted from those developed by aging at elevated temperatures. Although 50 years is probably not long enough for the strength of an alloy like 24S to attain that provided by aging at elevated temperature, eventually the strength on aging at room temperature should be greater. If a finite rate of precipitation obtains, I see no reason why room temperature should be any exception from the conventional behavior that the maximum strength or hardness and the time to attain it both increase with decreasing temperature of isothermal aging. Care must be exercised, however, to compare analagous portions of the aging curves; changes during the first few days of aging at room temperature may correspond to the initial hardness peak (as in the case of 24S alloy) whereas the main portion of the curve for elevated temperature aging corresponds to the second hardness peak.

J. B. HESS[†]—My comment involves a minor, and probably obvious point in connection with Dr. Geisler's description of the mechanism of "discontinuous precipitation," which he calls the "grain boundary reaction." According to the present viewpoint, this reaction is fundamentally recrystallization, with complete decomposition of the matrix accompanying the recrystallization. The author suggests that this recrystallization results from plastic strains produced by prior precipitation at the grain boundary areas.

A characteristic feature of the aging of magnesium base alloys is the onset of the grain boundary reaction at a very early stage of the process. In fact, metallographic and electron microscopic studies tempt one to believe that grain boundary precipitation in these alloys is of the pearlitic form, and therefore typically of the recrystallization type, *from the very start*. Based upon the hypothesis of prior precipitation, it is also difficult to rationalize the fact that the recrystallization reaction frequently proceeds from the grain boundary into only one of the adjoining grains.

Fortunately another source of the strains to initiate recrystallization suggests itself. Thermal contraction is anisotropic in magnesium, so that in polycrystalline samples this property should produce strains during cooling from the solution heat treatment, identical in character to those reported to produce fatigue by thermal cycles in tin, cadmium, and zinc. While no evidence of thermal fatigue in magnesium has been reported, so that the severity of such strains in magnesium are probably less, still I have been able to demonstrate their existence by use of Barrett's X ray microscopy technique.

The grain boundary reaction has also been reported to be almost the exclusive type of precipitation in tin alloys containing bismuth (J. E. Burke and C. W. Mason: *Trans.* AIME (1942) 147, 300) where thermal contraction strains should be more severe.

Accordingly, the mechanism of "discontinuous precipitation" should probably be generalized to include recrystallization resulting from strains other than those of prior precipitation.

A. H. GEISLER—Perhaps the precipitation strains are not always responsible for the recrystallization reaction but quenching strains in anistropic metals may sometimes contribute. On the other hand, these could hardly account for the reaction frequently observed with alloys of the isotropic metals Al, Cu, and Fe. In addition, I do not think that the existing data are sufficient to conclude that the recrystallization reaction can progress in the absence of general precipitation in any alloy system.

I appreciate the usefulness of the electron microscope in extending the resolution of precipitates but some of the existing replica techniques do not permit full utilization of this increased resolution. I would not conclude that general precipitation was absent without consulting X ray patterns of single crystals. While a minimum quantity in excess of a few percent by volume must be present in order to detect the precipitate by X ray diffraction, the restricting particle size may be well below that for electron microscopy as currently applied to metals.

A slight misstatement of my conclusions regarding the recrystallization reaction should probably be clarified. This process is presumed to result from plastic strains produced by precipitation throughout the grains, not at the grain boundaries alone. It is nucleated at grain boundaries not because of preferential precipitation here but because the activation energy is lower much as in the cases of localized precipitation and the decomposition of austenite into pearlite. I see no difficulty in rationalizing the morphology of the recrystallized nodules; it is identical with that of pearlite. Growth into only one of two adjoining grains suggests a favored orientation relationship between the distorted matrix and the recrystallized matrix analogous to a recrystallization texture.

F. C. HULL[*]—One of the previous discussers asked whether long time aging at room temperature would be comparable to a shorter period at an elevated temperature. I believe the results as judged from physical property tests might be very similar, but that if one could count the number and size of "coherent" regions and precipitated particles, a difference would be found. In the case of the isothermal transformation of austenite to pearlite at subcritical temperatures, it has been found that the rate of nucleation increases more rapidly with undercooling than the rate of growth.[†] At low transformation temperatures, therefore, the austenite reacts to form a larger number of smaller pearlite nodules. Nucleation and growth in age hardening systems might behave in a similar manner.

A. H. GEISLER—It is generally agreed that the rate of growth of the precipitate as controlled by rate of diffusion decreases with aging temperature whereas the rate of nucleation as controlled by both the rate of diffusion and degree of undercool goes through a maximum with decreasing temperature of isothermal precipitation. Thus, complete precipitation results in more numerous but smaller particles at the lower aging temperature, a condition which has been assumed to be responsible for the greater ultimate hardness and strength.

W. A. MUDGE[‡]—I do not think we should allow the chairman to close this meeting without an expression of gratitude for the excellent work which has been done in this case by Dr. Geisler. It is only through such careful work as this that we may hope to understand eventually what happens in the complicated commercial age-hardening alloys, particularly those of high melting point, such as the nickel-base material, where age hardening is accomplished by two or three elements. I hope that Dr. Geisler and his associates will at some future time have an opportunity to study these more complicated systems.

---

[*] Grumman Aircraft Engineering Corporation.
[†] University of Chicago.

[*] Westinghouse Research Laboratories.
[†] F. C. Hull, R. A. Colton, R. F. Mehl: Rates of Nucleation and Growth of Pearlite. *Trans.* AIME, (1942) 150, 185.
[‡] International Nickel Co.

A. H. GEISLER—While studies of the mechanism of precipitation have been confined in the past to binary alloys for simplicity, they are now being extended to multi-component alloys. The first step has been considered alloys in which only one phase precipitates but eventually it will be desirable to study alloys in which two or more phases can precipitate simultaneously.

R. A. RANDEBAUGH*—It has been suggested by Bain that quench hardening might be considered a special case of age hardening, the properties as quenched being compared to the maximum properties in age hardening alloys, quenched and subsequently aged to develop optimum precipitated particle size.

I wonder if the author has given any consideration to the application of his theories to quench hardenable alloys?

A. H. GEISLER—Yes, the concept of coherency and the characteristics associated with this state have been considered for other reactions in solid solutions. The work in progress is concerned with superlattice formation but the eutectoid reaction will be investigated also. Current consideration of the martensite reaction as a nucleation and growth process and recognition of the product as a coherent transition lattice will facilitate the analogy.

A. GUINIER†—It is obvious that in the regions where the concentration of foreign atoms becomes abnormally high, the matrix lattice is more or less disturbed.

On the other hand, the evidence indicates in many cases that it is not possible to describe the structure of these zones as the structure of the precipitate with slight changes due to the size of the nucleus and the coherency with the matrix lattice.

For instance, in Al-Cu alloys,[39] the streaks, visible on the X ray patterns (with monochromatic rays) in the early stages of aging, split, in a further stage, into diffuse spots definitely different from the spots of the θ' precipitate. That means that the structure of the "G.P. zones" cannot be considered as an imperfect θ' structure.

In Al-Ag or Al-Zn[40] alloys, the beginning of the aging is characterized by the fact that each node and also the center of the reciprocal lattice is surrounded by the same scattering zone which has the shape of a spherical shell. Such a diffraction effect is quite different from the one- or two-dimensional diffraction effects quoted by Dr. Geisler. This effect cannot be explained on the basis of the structure of the precipitate (Zn or $Al\text{-}Ag_2$); but it is very easily accounted for by the hypothesis of nuclei enriched in solute atoms and keeping without great change the structure of the matrix lattice.

Recent reports show that the zones of gathering of the solute atoms are by no means always plate-like, but that they also can be nearly spherical.[41]

A. H. GEISLER—The discussion so kindly submitted by Dr. Guinier is concerned with the interpretation of diffuse X ray diffraction effects, a subject that was not discussed in the present paper but was considered in detail in Ref 7[7]. Unfortunately, the latter was not yet in print when the present paper appeared. I entirely agree that the diffraction effects do not always suggest plate-like particles but change gradually as one, then two and finally three dimensions of the particle become adequate for sharp diffraction through normal growth at anisotropic rates. On the other hand, the evidence mentioned for Al-Cu alloys is no criterion for concluding that the structure causing the diffuse diffraction effects is different from the θ' structure. Results for an Al-Mg-Si alloy described in Ref 7 show that while spots in patterns for thin platelets do not coincide with those for large particles of precipitate, the configurations in reciprocal space determined by the analysis of not one but of many patterns are intimately related. In this case there is no question that the diffuse diffraction effects originate in a structure similar to that of the thick precipitate particles rather than the matrix. A possible explanation of the spherical zone of diffraction about matrix nodes of Al-Ag and Al-Zn alloys based upon structure of the precipitate is discussed in the cited reference.

References

7. *Acta Crystallographica,* Nov. 1948, **1** (5), 238–252.

39. *Jnl.* de Phys. VIII (1942) 124.

40. *Metaux et Corrosion* (1943) **18**, 209.

41. F. R. N. Nabarro: Symposium on Internal Stresses—Inst. of Metals —London (1947) p. 237.

---

* Georgia Institute of Technology.

† Laboratoire d'Essais du Conservatoire National des Arts et Metiers, Paris, France.

Factors Affecting the Tensile Notch Sensitivity of Magnesium Alloy Extrusions

By I. CORNET, Member AIME

DISCUSSION

(*P. R. Kosting and W. A. Johnson presiding*)

J. C. McDONALD*—The type of work presented here by Professor Cornet had its origin in field observations which seemed to indicate that magnesium was more notch sensitive than aluminum. As it was shown in our previous ASTM paper[11] and as confirmed by Professor Cornet's results, the static notch sensitivity of wrought magnesium is similar to that of wrought aluminum. As shown in our ASTM paper, the field observations noted above were actually a consequence of the greater toughness of the high strength wrought aluminum alloys compared to the wrought magnesium alloys.

---

* Dow Chemical Co., Midland, Mich.

[11] Amer. Soc. for Test. Materials: *Proc.* (1946) 46.

Effect of Grain Size on Tensile Strength, Elongation, and Endurance Limit of Deep Drawing Brass

By H. L. WALKER, Member AIME, and W. J. CRAIG

DISCUSSION

(*A. I. Blank and E. W. Palmer presiding*)

A. I. BLANK*—This paper is a valuable addition to the published data on the effect of grain size on endurance properties of deep drawing brass sheet. Previous investigators[7,8] of annealed wrought copper alloys have dealt mainly with the commercial ranges of grain size, and it is interesting to note that the data presented in this paper for commercial grain sizes closely confirm the earlier findings. This is probably the first time, however, that endurance properties for deep drawing brass of less than 0.010 mm grain size have been reported. Full advantage should be taken of the high endurance strength which is associated with these very small grain sizes.

As the authors have pointed out briefly, however, such grain sizes provide less ductility for forming purposes and, in addition, it will be appreciated that very small grain sizes are not too readily obtained in ordinary commercial practice. Special processing schedules and precise control of even nominal impurity content would normally be required to produce such grain sizes consistently and without the danger of incomplete recrystallization. These limitations are not intended, however, to detract from the merit of the authors' contribution.

H. L. WALKER (authors' reply)—The specimens upon which this work was conducted are from normal commercial production. The mill does not find it to be such a difficult task to control the grain size rather closely and the specifications call for a rather small variation in grain size. I should say the smaller grain sizes can be controlled within ± 0.002 mm diam. It is important to know the effects of ready-to-finish grain size and the amount of final cold reduction, and especially the latter. We have not encountered difficulties in securing complete recrystallization during the final anneal. The times and temperature of annealing are such that coalescence does not readily take place, which gives considerable flexibility in the annealing schedule. I should again like to point out that brass with the grain sizes reported here is being produced commercially.

L. W. GLEEKMAN†—I trust Professor Walker will not find me overbearing when I bring up a topic he said was not going to be developed in this paper. That was with regard to the effect of grain size on tensile strength and grain size.

I notice your Fig 3 does not go beyond the value of grain size of 0.024 mm diam. I was curious as to whether the curve of elongation vs. grain size would show a maximum elongation at an intermediate grain size. My reason for this is that work I did on a 14-carat gold alloy showed that the elongation rose to a maximum at a grain size of 0.025 mm and then fell off rather appreciably at higher grain sizes. I recall that Jeffries and Archer[9] and also D. K. Crampton,[10] showed similar results for elongation versus grain size.

I wonder if elongations for the two larger grain sizes were reported, would you get that maximum and subsequent decrease?

H. L. WALKER—Professor Gleekman is not overbearing in asking the question. We felt that the larger grain sizes with respect to tensile strength and elongation had already been reported in the literature a number of times. I should expect the elongation to reach a maximum value and then decrease with very large grain size. The tensile strength would not however reach a minimum value and then increase with increased grain size.

E. W. PALMER*—I have a couple of points that I would like to bring up. One of them is concerned with the fact that starting with an initially small grain size and giving fairly heavy reductions is a very fine way to develop directionality in a material. I am wondering if Professor Walker cares to comment on what influence directionality might have had on his work, especially with respect to the direction of test of a fatigue specimen.

There is another point. In explanation of some of the annealing curves, there is some discussion of a recovery period. If this starting material were stress relief annealed to start with, I am puzzled as to what you mean by "recovery period," in Fig 1.

H. L. WALKER—With respect to the production of directionality I wish to comment that we believe the rolling schedule affects the production of directionality. In the paper as originally written the rolling schedule was specified, but the reviewers declared it to be unimportant and it was deleted from the manuscript. The rolling schedule does affect directionality and stresses. We have found that some alloys prone to fire cracking during annealing may have the tendency to fire crack mitigated with proper adjustment of the rolling schedule. The diffraction patterns after recrystallization, and the tensile strength at 45 and 90° to the direction of rolling did not indicate any marked degree towards directional properties.

As to recovery, I consider recrystallization phenomena cover three distinct processes, the first called recovery, the second called recrystallization, and the third called coalescence.

Recovery I take to mean when in the X ray diffraction pattern there is a sharpening of the lines without appearance of any other visual changes. In other words, I believe there is a bringing together of the atoms into positions which give much sharper lines. Some may call this the incubation period for recrystallization. There may be the establishment of nuclei, but if so, they are not apparent from a diffraction pattern nor are they visible under the microscope.

The second process, recrystallization itself, I would define as the production of strain free grains through nucleation and growth, and strain free grains growing at the expense of strained grains for the establishment of an essentially strain free structure.

The third process, coalescence, is the growth of strain free grains at the expense of other strain free grains, and probably the driving force, or the energy exchange, is surface energy.

E. W. PALMER—With regard to recovery, would not that process of relief of strain on an atomic scale have occurred already in the strain relief anneal given to the material with which you started?

H. L. WALKER—Certainly some would have taken place. There is no question about that. But it does not necessarily indicate that it was complete, because I believe that recovery did take place in the series of anneals shown in Fig 1.

References

7. C. H. Greenall and G. R. Gohn: Fatigue Properties of Non-Ferrous Sheet Metals. *Proc.* Am. Soc. Testing Mats., (1937) **37**, Part II, 160.
8. H. L. Burghoff and A. I. Blank: Fatigue Properties of Some Coppers and Copper Alloys in Strip Form. Annual Meeting, June 1948, of Am. Soc. Testing Mats.
9. Jeffries and Archer: The Science of Metals. 1924, p. 161, McGraw-Hill, N. Y.
10. D. K. Crampton: *Metals Progress*, (1936) **29**, 39–42.

---

* Chase Brass and Copper Co.
† University of Delaware.
7,8 References are at the end of the discussion.

* American Brass Co.

A Copper-base Alloy Containing Iron as a High-strength High-conductivity Wire Material

By W. HODGE, R. I. JAFFEE, Junior Member AIME, J. G. DUNLEAVY and H. R. OGDEN

DISCUSSION

(A. I. Blank and E. W. Palmer presiding)

P. G. MAGANUS*—The type of work we do is in the resistance welding field, and we are wondering if this particular alloy might not be satisfactory in the making of these particular machines we have in mind. We would like to know what the magnetic properties of this particular alloy are. In other words, if put in the field on current, we would like to know whether or not it would heat up.

W. HODGE (authors' reply)—I cannot give a definite answer to your question. The alloy we know is ferromagnetic. As to applying it to the application about which you have asked, we have done no experimental work along that line.

L. PESSEL*—What are the fatigue properties of the alloy?

W. HODGE—We have actually made no fatigue tests on the alloy.

P. G. MAGANUS—I thought about melting and casting. I also happen to be a foundryman. I was wondering what this "Riser-X" was. The authors have it in quotes. Is that an iron base? I would imagine so, since it is exothermic.

H. R. OGDEN (authors' reply)—Yes, it is.

---

* Progressive Welder Co.

* Radio Corporation of America.

A High Strength-High Conductivity Copper-silver Alloy Wire

By W. HODGE, R. I. JAFFEE, Junior Member, J. G. DUNLEAVY and H. R. OGDEN, Junior Member AIME

DISCUSSION

(A. I. Blank and E. W. Palmer presiding)

W. HODGE—Since this paper was written the Signal Corps has sponsored the commercial development of this alloy. Work was carried out by the American Brass Co. under the able direction of R. D. Hull, and we are quite certain now that it is possible to produce an alloy of 6½ pct silver which will have high strength and high conductivity.

M. G. CORSON*—I am not in the habit of congratulating authors and I am quite sure they are in no need of congratulations, but I am glad that due to their efforts silver might become far more useful technically than it is today.

The characteristics of Cu-Ag alloys intrigued me long ago and it is now more than ten years since I broached the subject to the late W. H. Bassett of the American Brass Co. I did not experiment to such a great extent as the present authors did, nor did I get highly precise and uniform results. I saw, however, that with as little as 3 pct silver one might have copper conductor wire with over 100,000 psi strength and a conductivity of 90 pct. Besides, such wire is immune to permanent softening when exposed to a temperature up to 450°C.

Mr. Bassett could not see my suggestion from the commercial viewpoint, for obviously a wire with 3 pct silver would cost at least double the price of ordinary copper. It is a fact that OFHC copper in spite of all its advantages has hard sledding because its price is about one cent higher per pound. This cost question will form an enormous stumbling block in the way of Cu-Ag alloys, but I hope the future belongs to it.

By the way, I also suggested such uses of copper to the Mexican government, for that country could produce far more silver if there were a demand for it. My suggestion was not even honored with a reply. It is ridiculous that silver cannot have a definite price because it is obtained mostly as a byproduct, and the price of silver depends upon the arbitrarily fixed value of residues from copper, lead and zinc production. If we were not still under the spell of the idea that silver is a precious metal of an inherently high price, the latter might (and should) drop enormously and silver would become just a technical metal; that is where it belongs.

Regarding the degassing of Cu-Ag alloys I should suggest to the authors that they try blowing into the molten alloy first a stream of carbon monoxide in order to eliminate oxides and oxygen, then a stream of dry, oxygen-free nitrogen to remove whatever CO is retained. I found the method to work well in a number of cases.

P. J. MAGANUS*—I am in full agreement with Mr. Corson on the fact that these alloys are rather expensive, because we have worked with these alloys, too, in an experimental foundry that we have. If it were not for the fact that the stuff costs around 80 cents a pound, we probably could use it also in the resistance welding field.

One of the questions that I have is the fact that, in our foundry, when we used magnesium as a deoxidizer it also acted as a hardener all right, but it has the effect of absorbing gases or at least we get a lot of porosity when we have magnesium. When we forget to add magnesium and add other deoxidizers such as lithium or calcium borate or some of the other deoxidizers that work pretty well, we do not have that trouble. Can the author give us an idea?

In Fig 1 he gives the conductivity of these various elements and how much they depress the copper. I was wondering what magnesium does. In other words, I know that in the work we have done magnesium can knock down our conductivity to around 80 to 85 pct of copper, and with this conductivity we feel that we should not use it. Has the author some information on that?

W. HODGE (authors' reply)—The point you brought up, of porosity caused by magnesium, is very pertinent. In the laboratory we did not have trouble from porosity caused by magnesium, presumably because of the small melt.

When we attempted to cast larger ingots for cold rolling, the porosity caused by magnesium ruined our entire first attempt, and Mr. Hull, of the American Brass Co., finally decided to throw some borax on top of the melt, and by so doing he was able to get rid of most of the porosity caused by magnesium. Otherwise we could not have cast the alloy in the air at all. It would have had to be cast quite in a protected atmosphere.

---

* Consulting Engineer, New York City.

* Progressive]Welder;Co.

On the other point on conductivity, I could only refer you to a rather old article by M. G. Corson, in which he described the effect of magnesium copper alloys. I think it was written in about 1930 and occurs in the AIME *Transactions*.

E. W. PALMER*—I should like to congratulate the authors on the very fine development work that has resulted in these two new high-strength high-conductivity alloys.

The situation in copper-silver and copper-iron was gone over carefully years ago, looking for precipitation hardening effects. When no hardening was discovered, the systems were dropped.

The discovery by Mr. Hodge and his co-workers, that these precipitation hardened matrices have enormously high rates of work hardening that permit them to be drawn to very high strength is of fundamental importance, and the authors deserve a great deal of credit for this development.

---

* American Brass Co.

The Effect of Chromium on the Ms Point

By J. B. BASSETT and E. S. ROWLAND, Junior Member AIME

DISCUSSION

(*C. S. Smith and A. R. Troiano presiding*)

A. E. NEHRENBERG*—From the practical viewpoint it is not particularly important whether 1 pct chromium lowers the *Ms* temperature of steels containing 1 pct carbon 65°F as indicated by the data just presented, or about 40°F as shown by Klier and Troiano.[2] This is because such high carbon steels are not austenitized commercially at a high enough temperature to dissolve all of the carbides present. In such cases the actual *Ms* temperature is considerably higher than one would predict on the basis of the known effects of the alloying elements present. However, it is somewhat disturbing that there should be this disagreement among these investigators. Perhaps a few comments on this discrepancy would be in order.

It seems to this discusser, in spite of the arguments of the authors to the contrary, that the possibility exists that the time allowed for the hot oil quench used for temperatures below 350°F may not have been long enough to permit the specimens to reach the bath temperature before they were tempered. It may be significant that in the authors' Fig 6 which shows the relation between the *Ms* point and chromium content for the two carbon levels studied, the five points which do not agree with published data all were obtained using hot oil for the initial quench. For the compositions which have *Ms* temperatures high enough so that molten salt could be used for their determination, there is good agreement with published data, even at the 1 pct carbon level. It is recognized, however, that there is one point based on a hot oil quench which does not deviate from the expected value.

In their paper Klier and Troiano[2] indicated that steps were taken to establish that the temperatures of the quenching baths used were attained. There is no indication that this was done in the investigation just described. Therefore this discusser is inclined to agree with Klier and Troiano that 1 pct chromium lowers the *Ms* temperature of steels no more than about 40°F irrespective of variations in carbon from about 0.40 to 1.3 pct.

It is of interest to compare the *Ms* points reported in the present paper with calculated values. Formulas which have been proposed for calculating *Ms* temperatures contain factors for carbon which vary from 540 to 650.[1,8,9] The significance of this variation in the carbon factor can be established if common factors are assigned for each of the alloying elements in these formulas other than carbon.

The *Ms* temperatures shown in Table 6 were calculated by using the Grange-Stewart and the modified Payson-Savage formulas with common factors of 70 and 40 for Mn and Cr, respectively. The effect of the residual Si and Ni was neglected.

It will be noted that when the carbon content is about 0.50 pct the difference in the carbon factor used in these formulas has no effect on the accuracy of the calculated *Ms* temperature, and that there is good agreement with the measured values. The maximum deviation is only 20°F. When the formula containing the higher carbon factor is used 9 of the 12 steels have calculated *Ms* points which differ by only 10°F or less from the measured values. When the formula containing the lower carbon factor is used the calculated *Ms* temperatures of 10 of the 12 steels differ by no more than 10°F from the measured values.

At the 1.0 pct carbon level, however, calculations based on the formula having the higher carbon factor tend to be considerably lower than those based on the one with the lower carbon factor, as the table indicates, and do not agree as well with the measured values.

Earlier in this discussion it was suggested that the *Ms* temperatures shown for the five higher chromium steels might be low. If that is so, there would be even better agreement between calculated and measured *Ms* points for the modified Payson-Savage formula than the table indicates and a greater deviation for the Grange-Stewart formula. Be that as it may, the results as they stand indicate that the modified Payson-Savage which contains the lower carbon factor appears to be more satisfactory at this high carbon level. Additional data which confirm this fact were presented in my discussion of the Grange-Stewart paper.

The carbon factor used in the Grange-Stewart formula is based on data published by Greninger[10] and by Digges[11] which showed that 1 pct carbon in pure iron-carbon alloys lowers the *Ms* temperature 650°F. The carbon factor in the modified Payson-Savage formula was also based on data published by Greninger[10] which showed that 1 pct carbon in commercial iron-carbon alloys corrected to 0 pct Mn lowers the *Ms* temperature 540°F. Since these differences in the carbon factor were reported by a single investigator it seems reasonable to conclude that the impurities present in carbon steels have an effect on the rate of lowering of the *Ms* temperature with increasing carbon.

Grange and Stewart in the closure to their paper[1] also presented data to show the effect of variations in carbon on the *Ms* temperature of carbon steels. These data and those of Greninger, corrected to 0 pct Mn, are plotted in Fig 7. There is good agreement among these investigators. Grange and Stewart not only confirm that 1 pct carbon lowers the *Ms* temperature of commercial steel 540°F, but also show that the effect of carbon is not linear above about 1 pct. The rate of lowering of the *Ms* temperature with increasing carbon above 1 pct gradually decreases so that a correction based on Fig 7 must be made if the *Ms* temperature for a steel containing more than this amount of carbon is to be calculated.

---

* Crucible Steel Co. of America.

[2] References are at the end of the discussion.

It cannot be emphasized too strongly that *Ms* temperatures can be calculated only when the austenitizing temperature is sufficiently high so that there are no residual carbides present. Since this generally occurs commercially in only the medium carbon steels where either of the formulas discussed above is equally accurate, it usually makes little difference whether one or the other is used for the calculation of the *Ms* temperature. However, it should be remembered that the effect of carbon in commercial steels is somewhat less than the Grange-Stewart formula implies.

If it is desired to calculate the *Ms* temperatures of steels containing 1 pct carbon only the modified Payson-Savage formula can be used to obtain values which will be reasonably accurate.

The authors conclude that "as far as steels containing chromium are concerned, no simple formula is adequate for the accurate calculation of *Ms* points." This is contradicted by the fact that there is reasonable agreement with the authors' own data when the *Ms* temperatures are calculated using the modified Payson-Savage formula. Furthermore, data presented in my discussion of the Grange-Stewart paper showed that the agreement between calculated and measured *Ms* temperatures was remarkably good for the chromium steels 5140 and 52100 and for the 1 pct carbon steels containing from 1 to 9 pct chromium studied by Klier and Troiano.

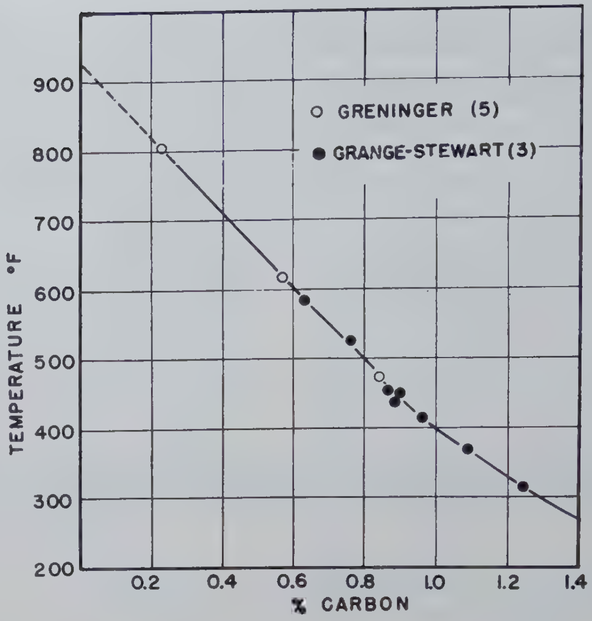


FIG 7—Effect of carbon on *Ms* temperature of carbon steels. Variations of carbon in plain carbon steels between about 0.2 and 1 pct lower the *Ms* temperature at the rate of 540°F per 1 pct carbon.

A. R. TROIANO*—There seems to be some question about the disagreement in the rate of lowering of *Ms* with chromium in the high carbon steels. Did the authors make any grain size determinations, or could they tell us approximately what the grain size was? In the work that Dr. Klier and I did, we had exceptionally large grains, and since we now know that the grain size may change *Ms* by as much as 35°C, it is possible that the disagreement may be caused by this phenomenon. If the authors had a smaller grain size, the deviation would be in the right direction.

E. S. ROWLAND (authors' reply)—The authors are indebted to the discussers of this paper for their comments and criticisms.

First, in answer to Mr. Nehrenberg, we fear that he missed the implication contained in the opening paragraph of the paper in regard to the effect of small variations in chromium content on the *Ms* temperature. The curve of Fig 11 of the Grange and Stewart[1] paper indicates that the effect of chromium on *Ms* might be very much greater than previously anticipated in the range of chromium contents normally expected to be in solution in the austenite of SAE 52100 at its commercial austenitizing temperature (about 1550°F). It was conceived possible that small variations in austenitizing conditions might produce a significant shift in the martensite temperature range if the above curve expressed the true relationship. Mr. Nehrenberg is right in stating that the difference between 40 and 65°F per 1 pct chromium is unimportant but an effect two or three times either of these values at low chromium contents would have been important. As it turned out, the relationship between chromium and *Ms* temperature is undoubtedly linear and this effectively disposes of any such arguments.

Mr. Nehrenberg's second criticism is much more serious since it throws doubt upon the validity of all the data obtained by questioning the fact that the specimens actually reached the quenching bath temperature. In the first place, the

Table 6 . . . Effect of Carbon Factor on Calculated Ms Temperatures

Steel	C	Mn	Si	Ni	Cr	Measured Ms, °F	Calculated Ms		Deviation	
							540 (*a*) Carbon Factor, °F	650 (*b*) Carbon Factor, °F	540 Carbon Factor	650 Carbon Factor
1	0.53	0.79	0.30	0.12	0.01	590	590	600	0	+10
2	0.505	0.79	0.29	0.12	0.24	586	590	605	+4	+19
3 (1)	0.51	0.79	0.30	0.11	0.51	575	580	595	+5	+20
(2)	0.46	0.80	0.355	0.14	0.56	610	605	625	−5	+15
4	0.51	0.80	0.28	0.12	0.74	575	570	585	−5	+10
5	0.505	0.79	0.29	0.12	1.03	570	560	575	−10	+5
6	0.495	0.77	0.29	0.12	1.27	565	555	570	−10	+5
7	0.50	0.80	0.28	0.12	1.54	555	545	560	−10	+5
8	0.47	0.80	0.355	0.21	2.09	555	535	555	−20	0
17	0.485	0.34	0.24	0.11	1.05	620	600	620	−20	0
18	0.48	0.35	0.25	0.11	1.32	600	595	610	−5	+10
19	0.48	0.34	0.26	0.13	1.45	595	590	605	−5	+10
9	1.075	0.33	0.30	0.12	0.01	330	325	275	−5	−55
10	0.95	0.35	0.24	0.11	0.27	400	380	350	−20	−50
11 (1)	0.965	0.35	0.255	0.10	0.53	380	365	330	−15	−50
(2)	0.965	0.35	0.34	0.12	0.55	365	365	330	0	−35
12	0.97	0.35	0.26	0.11	0.78	335	350	315	+15	−20
13	0.97	0.35	0.24	0.12	1.04	330	340	305	+10	−25
14	0.965	0.35	0.28	0.12	1.28	313	335	300	+22	−13
15	0.96	0.36	0.265	0.12	1.54	300	325	290	+25	−10
16	0.97	0.37	0.35	0.13	2.05	275	300	265	+25	−10

(*a*) Ms (°F) = 930 − 540 C − 70 Mn − 40 Cr (Based on Payson-Savage Formula)
(*b*) Ms (°F) = 1000 − 650 C − 70 Mn − 40 Cr (Based on Grange-Stewart Formula)

---

* University of Notre Dame.

specimens used were rough cut to the sizes indicated in the text but by the time they were cleaned up by hand grinding, the actual sizes were about $\frac{3}{32}$ in. square and $\frac{5}{32}$ in. square, respectively. Nevertheless, actual cooling times were obtained on $\frac{1}{8} \times \frac{1}{8} \times \frac{1}{2}$ in. long specimens, after austenitizing at 1800°F, during a salt quench at 350°F. This work was done by welding 24 ga thermocouples to each of two specimens and connecting them to two channels of a Hathaway 12 channel recording oscillograph. The strings had a dc sensitivity of 55 mm per mil amp at 1 meter distance and a natural frequency of 400 cycles per sec. The results showed that both specimens were within 10°F (about the experimental error of the determination) of the bath temperature in 4 sec.

At several times during the investigation, the smaller specimens were checked against the larger, using the austenitizing conditions and quenching times in both salt and oil described for the high carbon steels. In all cases the results were identical. Furthermore, both oil and salt were used interchangeably in the range of 300 to 350°F with the same results on many check runs. Finally, duplicate specimens quenched together using the oscillograph produced differences in temperature of 25°F when only 50°F above the bath temperature and separate runs showed even greater deviations. It is believed that these data support the contention that reproducibility of results is a satisfactory indication that the specimens have actually reached the quenching bath temperature. So many variables enter into the actual quenching operation that reproducibility is out of the question if the specimen is removed from the bath while still above the bath temperature.

Mr. Nehrenberg also criticizes our statement that no simple formula will suffice to express the relationship between the effects of carbon and chromium on the *Ms* temperature and in defense of his objection shows that the modified Payson-Savage formula will predict the *Ms* points within 25°F. Our statement was based on the fact that the effects of carbon and chromium on the *Ms* points of these steels are interdependent and not independent as formulas of the Payson-Savage and Grange-Stewart type assume. Either or both may arrive at a reasonable prediction by a series of compensating errors, but a more complex formula would obviously be required to express accurately the true relationship between carbon and chromium in this regard.

In answer to Dr. Troiano's question regarding the grain size of the specimens after the austenitizing treatment, samples of 1.5 pct chromium steel were heated for 20 min. at 2100°F and isothermally transformed at 1300°F for 3 min. to delineate the grain size. The resulting ASTM grain size was No. 2. We are unable to assess this result in terms of Dr. Troiano's statement of "exceptionally large grains" obtained in his work.

Since the manuscript for this paper was submitted, both Dr. Troiano and Dr. Cohen have shown an effect of grain size on *Ms*. Since high austenitizing temperatures are necessary for the complete solution of high carbon chromium steels, coarse grain size is unavoidable and its effect on *Ms* difficult to separate from carbide solution since both are changing simultaneously. Our only pertinent observation in this regard was that with increasing austenitizing temperature, the *Ms* became constant shortly after the complete disappearance of carbide in the quenched microstructure and did not change upon a further increase of austenitizing temperature of 100°F.

References

1. R. A. Grange and H. M. Stewart: The Temperature Range of Martensite Formation. *Trans.* AIME (1946) **167**, 467–501.
2. E. P. Klier and A. R. Troiano: Ar″ in Chromium Steels. *Trans.* AIME (1945) **162**, 175–185.
8. E. S. Rowland and S. R. Lyle: The Application of *Ms* Points to Case Depth Measurement. ASM *Trans.* (1946) 37, 27–47.
9. A. E. Nehrenberg: Discussion of Ref. 3.
10. A. B. Greninger: The Martensite Thermal Arrest in Iron-Carbon Alloys and Plain Carbon Steels. ASM *Trans.* (1942) **30**, 1–26.
11. T. G. Digges: Transformation of Austenite on Quenching High Purity Iron-Carbon Alloys. ASM *Trans.* (1940) **28**, 575–607.

The Kappa Eutectoid Transformation in the Copper-silicon System

By W. R. HIBBARD, JR., Junior Member, G. H. EICHELMAN, JR., Student Associate AIME, and W. P. SAUNDERS

DISCUSSION

(C. S. Smith and A. R. Troiano presiding)

C. S. SMITH*—The authors refer to some earlier work of mine containing reference to the pearlitic structure of these alloys. It should, perhaps, be pointed out that the two-phase structure most reminiscent of pearlite is formed when gamma precipitates from supersaturated alpha phase or from an alpha-kappa mixture rich in alpha. This is not a eutectoid decomposition involving three different phases, but is a case of discontinuous precipitation involving two phases (one, however, possibly of two different compositions). It seems not unlikely that a typical nearly-parallel-plate form of pearlite occurs only when there are three interfaces of approximately equal energy meeting at the advancing boundary of the transformed alloy. When alpha forms from kappa in an oriented manner, the resulting interface is of extremely low energy and the gamma phase can co-precipitate with alpha without the constraint of the third interface.

C. S. BARRETT*—Dr. Smith recently gave me a set of samples of copper-silicon alloys that I have been X raying after different heat treatments. The investigation is not yet completed, but it is already clear that the decomposition of a sample of homogeneous kappa into alpha plus gamma on cooling below the eutectoid temperature is crystallographically very different from the precipitation of alpha from kappa on heating above 600°C. The diffraction patterns in the two cases are as strikingly different as the appearance of the microstructures, and in the way one might expect from the microstructures: there are normal reciprocal lattice points for all phases after the eutectoid transformation, but after the precipitation, which produces the remarkable striations in the microstructure that Dr. Smith has reported,[5] reciprocal lattice points are often seen that are greatly elongated. Elongated points are found after only one minute at a precipitation temperature of 750°C but none is found after times up to 45 hr at 500°C.

A. R. TROIANO*—Dr. Barrett has

---

* University of Chicago.

* University of Chicago.

* University of Notre Dame.

raised an interesting point. Care must be taken in generalizing the term "pearlite" if one is going to draw the analogy with steel. One may have very pearlitic-looking microstructures that do not originate from the eutectoid decomposition of the parent phase.

W. R. HIBBARD, JR. (authors' reply)—Dr. Smith has emphasized a point which the authors failed to realize, perhaps because it is somewhat difficult to understand how kappa, even in an alpha-kappa structure, can decompose to plate-like alpha and gamma without involving the eutectoid transformation (see Fig 20 of Ref. 5).

Dr. Troiano's caution regarding the use of the term "pearlite" is most timely. The striated structure of alpha and kappa formed above 600°C from kappa described by Dr. Barrett obviously is not a eutectoid structure, since gamma is not involved, and it probably looks more "twin"-like than "pearlite"-like, anyway.

Stabilization of the Austenite-martensite Transformation

By W. J. HARRIS, JR., Junior Member, and M. COHEN, Member AIME

DISCUSSION

(C. S. Smith and A. R. Troiano presiding)

A. R. TROIANO[*]—This paper is full of new observations.

It is very interesting to me to see that now we can rationalize why in some cases one may observe stabilization without any decomposition at all. I distinctly remember how in the laboratories at Notre Dame we found this phenomenon and then tried to check it in other steels and it did not seem to be there. We were beginning to wonder if the original determinations were valid. Now we can see good reasons why it was not observed in all steels.

I wonder if the authors would care to state just how general the relationship between σ_s, and M_s might be.

C. S. BARRETT[†]—Our investigation of the transformation in lithium and lithium-magnesium alloys, which is somewhat like the martensite transformation in steels,[14] is continuing. We are using an improved X ray diffraction unit that has been built by D. F. Clifton to see whether there is stabilization of the high temperature phase on cooling, and to investigate further a behavior that looked like stabilization of the low temperature form during reversion to the high temperature form on heating. Detailed information of this kind for the lithium transformation should be quite helpful in arriving at sound theories of stabilization.

B. S. LEMENT[‡]—I just want to point out that even though stabilization of austenite can result from quenching to and holding at some temperature below σ_s, the extra austenite at room temperature is not stabilized against room temperature transformation. The reason for the apparent contradiction has been pointed out by Dr. Morris Cohen in his Campbell Memorial Lecture. There are two different mechanisms involved in the transformation of austenite to martensite. On cooling the transformation probably occurs by a process of nucleation and shear; whereas the isothermal transformation at room temperature probably occurs by growth of the existing martensite plates. Any extra austenite that results from stabilization on cooling to room temperature behaves quite similarly to the other retained austenite with regard to isothermal transformation at room temperature. If more austenite is retained, more transformation occurs on aging at room temperature. This phenomenon is of great importance from the standpoint of dimensional stability of tool steels.

L. D. JAFFE[*]—The authors have certainly done a fine job in clarifying and establishing many points only suggested by previous work. On one matter, however, there seems to be some disagreement. The authors' Eq 1 indicates that for constant percent martensite (P), M_s-T is a constant. In other words, all curves of percent martensite vs. temperature are the same, except for a uniform shift along the temperature axis. This is not in accord with the data of Grange and Stewart,[25] obtained by technique similar to that of authors. As the writer has pointed out,[26] these data indicate that M_s-T increases as M_s decreases. Can the authors throw any light upon these apparently contradictory findings?

It may be noted that the discrepancy just mentioned does not affect the authors' conclusion that all the (unstabilized) martensite transformation curves conform to a single function, in which M_s is the only metallurgical variable. This same conclusion was, indeed, drawn by the writer in the discussion cited.[26]

M. COHEN (authors' reply)—The σ_s relationships that Dr. Troiano has inquired about seem to hold for the following ranges of composition: 0.75 to 1.35 pct carbon, 0–5½ pct nickel, and 0–3 pct chromium. These limits are not definitely fixed and may be extended when future work is done on additional steels. However, considerable deviation may be expected when the composition is changed radically, as in the case of high speed steel.

We agree with Dr. Barrett that his contemplated studies on lithium and lithium-magnesium alloys will have a significant bearing on the theory of stabilization. For example, in working with steels, we are inclined to wonder about the role of carbon in the stabilization process, whereas if true stabilization occurs in a high-purity metal, one may well question the importance of solute elements in the phenomenon.

Mr. Lement is entirely correct in emphasizing that the term stabilization used in this paper refers only to the inhibition of the cooling austenite-martensite transformation. The increment of retained austenite resulting from stabilization is subject to isothermal transformation in the same sense as normal quantity of austenite retained in the absence of stabilization. In fact, this is one of the main reasons for suspecting that the isothermal transformation of austenite to martensite[27] may occur, at least in part, by the growth of existing plates, rather than by the formation of new ones as in the cooling transformation.

It is worth repeating here that the σ_s temperatures determined in this paper are all based on holding times of ½ hr. At the moment, the relation between σ_s and holding time is not known. However, regardless of this possible variation, the main trends should persist, with stabilization becoming more pronounced, the larger the amount of martensite present.

Dr. Jaffe rightly points out that discrepancies exist between the martensite transformation curves presented here and those of Grange and Stewart. No positive explanation is available at the moment, although it should be pointed out that the method used by the latter investigators was only semi-quantitative. The form of the transformation curves as given by Eq 1 is identical with the one found by Howard[2] for carbon contents ranging from 0.75 to 1.35 pct. On the other hand, Grange and Stewart covered a wider range of carbon contents, and their data may indicate trends that are not detectable in a more limited series of compositions. However, according to our experience, quantitative measurements are quite difficult in the lower carbon steels and may lead to questionable results even when lineal analysis is employed.

---

[*] University of Notre Dame.
[†] University of Chicago.
[‡] Massachusetts Institute of Technology.

[*] Watertown Arsenal Laboratory.
[14] References are at the end of discussion.

References

R. T. Howard and M. Cohen: Austenite Transformation Above and Within the Martensite Range. *Met. Tech.* Sept. 1947, TP 2283.

14. C. S. Barrett: Low Temperature Transformations in Lithium and Lithium-magnesium Alloys. *Met. Tech.* (April 1948) TP 2346.
25. R. A. Grange and H. M. Stewart: The Temperature Range of Martensite Formation. AIME *Trans.* **167** (1946) 467–490.
26. L. D. Jaffe: Discussion of Ref. 25. AIME *Trans.* **167** (1946) 493–494.
27. B. L. Averbach and M. Cohen: The Isothermal Transformation of Martensite and Retained Austenite. A.S.M. Preprint No. 1 (1948).

Secondary Hardening of Tempered Martensitic Alloy Steel

By **W. CRAFTS**, Member AIME, and **J. L. LAMONT**

DISCUSSION

(C. S. Smith and A. R. Troiano presiding)

R. BALLUFFI* and M. COHEN*—In the tempering research conducted at M.I.T., carbide precipitation from martensite or retained austenite has never been found to produce an expansion. Consequently, the expansion effects of Fig 11 and 12, attributed by the authors to alloy carbide precipitation, are of some concern to us. We have observed such expansions, but have accounted for them on the basis of retained austenite decomposition. The authors have not used this explanation because they assumed that the retained austenite in their samples had been reduced to negligible amounts by previous refrigeration in liquid oxygen.

With the cooperation of the authors, we undertook a critical experiment to settle this issue. The authors kindly sent us four of the alloy steels under consideration, of which we selected the following two for test:

Designation	Heat	Pct C	Pct Mn	Pct Si	Pct Cr	Pct V
Cr	4–24	0.52	0.37	0.19	11.73	
V	4–257	0.48	0.54	0.12		2.02

Two dilatometric specimens of each composition were austenitized at 2300°F (1260°C) for 12 min.† and quenched in oil. One of each was immediately refrigerated in liquid nitrogen. All four specimens were then subjected to dilatometric runs, covering the significant range of 600–1300°F in 2 hr. The results are plotted in Fig 22 in a manner analogous to that of Fig 11.

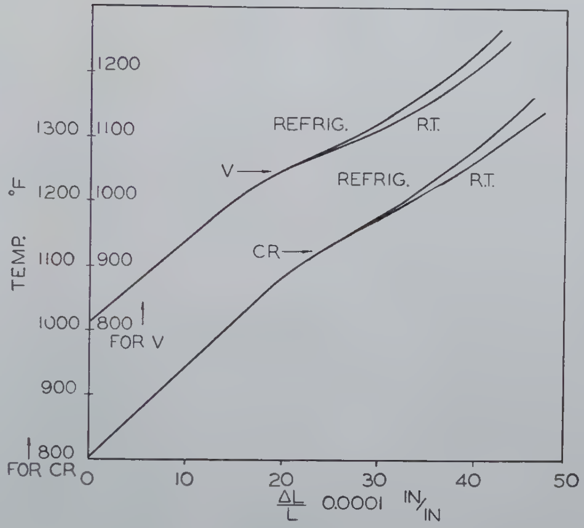


FIG 22—Dilation on heating 0.5 pct carbon, chromium and vanadium steels, after oil quenching from 2300°F and after refrigeration in liquid nitrogen.

The expansion effect is definitely smaller for the refrigerated steels than for the as-hardened steels. This clearly suggests that the expansion phenomenon is the result of retained austenite decomposition inasmuch as the refrigerated specimens undoubtedly contain less retained austenite than the as-hardened ones. If the expansion during tempering were caused by carbide precipitation from the martensite, the refrigerated specimens should exhibit at least as much expansion as the as-hardened ones, and probably slightly more because of the martensite increment produced by the cold treatment. It is also worthy of note that the tempering expansion of the refrigerated specimens represents a sizable fraction of the expansion of the as-hardened specimens. This means that the cold treatment does not convert anything like all of the retained austenite.

These findings should not reflect upon the authors' conclusions as to the cause of secondary hardening in the 0.50 pct carbon steels under study here. While the retained austenite decomposition should not be ignored, we quite agree that alloy carbide precipitation plays an important role in the secondary hardening observed.

D. J. BLICKWEDE*—I would like to point out to the authors two phenomena which occur during the isolation of carbides from steels that might affect their electron micrographs.

One is that carbon in the matrix does not ionize during electrolysis. Instead it either comes down as free amorphous carbon or reacts with hydrogen to form hydrocarbon gases. The presence of free amorphous carbon might affect the electron micrographs.

---

* Massachusetts Institute of Technology.
† Time at temperature.

* Naval Research Laboratory, Washington, D.C.

The second phenomenon occurs in the case of alloy steels, in which particles of the matrix material—that is, the steel itself—remain undissolved and will come down with the carbide precipitate. Thus it is quite likely that pieces of matrix material itself could be the carbide residue and show up in the electron micrographs.

W. CRAFTS (authors' reply)—The examination of isothermally transformed bainite should be very informative. A preliminary survey of bainite specimens has indicated that the carbide structures are exceedingly fine and probably quite complex, so that a good deal of difficulty in experimental technique and interpretation may be anticipated. It is expected however, that electrolytic extraction of the carbides may provide a useful tool for furthering our understanding of the nature of bainite. The carbide of a 3 pct chromium steel in the bainitic condition was found to change on tempering from Fe_3C to Cr_7C_3 in about the same temperature range as steel in the martensitic condition, but the specimens were not examined to determine the appearance of the carbides during the transition. In view of the apparently complex nature of the carbides in bainite, it might be expected that the transition might not be as simple and straightforward as in tempered martensitic steels.

In reference to the discussion by Robert Balluffi and Morris Cohen it should be noted that the analyses given in their discussion represent the same steels that are shown in Fig 11. By an oversight these specimens were indicated in the text to have come from the same heats as those used for the metallographic investigation. The dilation tests indicate that the larger amount of residual austenite in the unrefrigerated specimens increased the amount of dilation on heating in the 1000–1300°F range. It is also quite probable that the refrigerated specimens contained an amount of residual austenite that was too small to be detected in our tests. However, it is felt that a more positive test should be made before eliminating the possibility that the expansion is in part due to transition of the carbide phase. As is pointed out, the actual amount of retained austenite is too small to affect significantly the conclusions regarding the effect of transition to alloy carbide on the rehardening phenomenon.

In electrolytic extraction, as stated by D. J. Blickwede, a certain amount of carbide is decomposed to hydrocarbons and amorphous carbon. The attack occurs largely during electrolytic solution of the metal and, although subsequent solution by the electrolyte has been substantially eliminated, the recovery of carbide is never complete and in slightly tempered specimens may be quite low. Thus, the carbide residue is only a partial representation of the total amount and individual carbide particles may be altered from their original condition. The X ray indications, therefore, may be incomplete, but positive results may be accepted, and the appearance of the carbides parallels the collateral properties sufficiently well to give a credible confirmation that the observed structures are significant. Relatively little difficulty due to undissolved particles of matrix material has been encountered in steels that are free from large amounts of residual austenite and the presence of metallic iron in the residue has been found to be a useful criterion of retained austenite. Undissolved carbides and matrix material are too coarse to present any difficulty in the electron microscope examination. Although electrolytic extraction requires careful experimental technique, it offers a means of examining carbide structures that is much more revealing than methods of examination suitable for solid specimens.

Furnaces for the distillation of nitric acid. (Ercker, 1574.)

The tall tower was filled with charcoal which burned only at the bottom and the hot gases passed through the side chambers to heat the alembics containing vitriol, saltpeter and other materials. Fresh charcoal descended to replace that burned, and the furnace would maintain constant heat for as long as a day without refueling or other attention. (*Courtesy C. S. Smith*)

The Effect of High Copper Content on the Operation of a Lead Blast Furnace, and Treatment of the Copper and Lead Produced

A. A. COLLINS,* Member AIME

When we speak of high copper on a lead blast furnace we think in terms of 4 to 5 pct, or any lead charge carrying over 1 pct. Any copper on charge will produce its corresponding troubles such as lead well, extra slag losses, drossing problems, and the working up of the dross.

This is indeed a very interesting subject and one that used to give the old-time lead metallurgists such as Eiler, Hahn and Iles many worries, not so much in the actual operation of the blast furnace but in the working up of the copper. When the American metallurgists commenced with the American rectangular-shaped lead blast furnace in the 1870's and got away from the reverberatories such as were in use in Germany and other parts of the world, they went to greater tonnages, as 80 to 100 tons per day in comparison to the 20 to 30 tons per day in the other processes.

With the greater tonnages along with insufficient settling capacity, the silver losses in some cases were increased. Hence the lead-fall was low, for there were no leady concentrates in those days to assist the metallurgist to gain lead or an absorber for the *precious* metals; and in some cases copper sulphides were added intentionally to the charge to produce a copper matte to lessen the silver losses through the dump slag. The operators in those days thought that where some copper was always present in the lead ores the copper should not enter into the reduced lead and alloy with it. This, by the way, is just the reverse of our present-day practice, when we try to put all of the copper into the blast furnace lead and to remove the same through the drossing kettles. Therefore the furnace was operated to produce a certain amount of matte or artificial sulphides, since, due to the great affinity of copper for sulphur, any copper present would enter the matte almost completely. Thus, the lead bullion produced was practically free from copper.

The products of the furnace were metallic lead or lead bullion, containing 85 to 95 pct of the lead and about 96 pct of the silver which were in the ore—a lead-copper-iron matte which contained nearly all the copper in the ore and the slag, the waste product.

In the United States, up through the year 1892, we find the small furnace 100 × 32½ in. with 12 tuyeres, some 6 on each side, plagued with a small amount of poorly roasted sulphides—either from heap or hand roasters that produced matte.

This matte was roasted and if poor in copper was returned for the ore smelting. Otherwise it was smelted either alone or with additions of rich slags or argentiferous copper ores, the products being lead and a highly cupriferous matte, the latter being subsequently worked up for its copper.

The lead metallurgists kept trying and improving on furnace and roasting equipment designs until we find Malvin W. Iles constructing at the old Globe Plant at Denver what came to be the modern furnace. That is, in 1900 he built a furnace of 42 in. width by 140 in. at the tuyeres with a 10 in. bosh and a 16-ft ore column. This type has been more or less standard to the present time, though modified in width and length to meet the demand for large tonnages and improvements in structural details. In 1905 at Cananea, Mexico, Dwight and Lloyd developed the present down-draft sinter machine that has meant so much in producing a well-processed material for the lead blast furnace.

In 1912 Guy C. Riddell came forth with double roasting at the East Helena Plant of the American Smelting and Refining Co., which removed the "zinc mush plague." Incidentally, with the introduction of double roasting, which most lead plants were forced into after 1924, when lead flotation came into its own, less matte or no matte was produced.

When this stage arrived, the copper was forced into the dross and the casting of lead at the blast furnace lead-wells was stopped. In plants with a fair copper carry 1 pct or better on the blast furnace charge, the lead wells became inoperative once the production of matte stopped. The copper drosses clogged the lead wells and even with bombing, either water or dynamite, the operators could not keep them open. Thus, the lead wells were abandoned in some plants, such as at the El Paso and Chihuahua smelters of the American Smelting and Refining Co., and all lead taken out through the first settlers.

The elimination of sulphur, especially sulphide sulphur, from the blast furnace charge and the nonproduction of matte resulted in a great saving of time, energy and equipment in the recirculation of the copper.

With the copper content in the dross and dross-fall ranging in quantities from a few percent up to 60 pct, such as at El Paso, a drossing problem was created.

As the old-time operators hated dross and buried the same in the shipping bullion, the modern metallurgists from 1925 on decided that with increasing dross-falls they would have to adopt the lead refiner's ideas of drossing kettles with subsequent treatment of the lead with a sulphur addition to have the shipping lead of 0.01

San Francisco Meeting, February 1949.

TP 2576 D. Discussion of this paper (2 copies) may be sent to *Transactions* AIME before July 1, 1949. Manuscript received Nov. 1, 1948.

* Manager, Chihuahua Smelter, Chihuahua, Chih., Mexico.

pct copper content or less.

With the cleaning up of the bullion and thus the copper remaining in the lead smelters, the dross reverberatory was developed at the Selby Plant of the American Smelting and Refining Co. and brought to its present stage of development by Mr. O. P. Chisholm, the former superintendent of the East Helena Smelter.

Since 1925, when the real flotation of lead sulphide commenced, the concentrators have gone in some cases from a 10 to a 64 pct lead in the lead concentrates, and the copper has also increased in the same concentrates, in some cases up to 14 pct. Consequently, the lead smelters have received a considerable amount of copper in these lead concentrates.

These higher copper-lead concentrates along with complex copper, zinc, lead ores after roasting give a high copper charge to the lead blast furnaces, say up to 5 pct, and have in no way retarded the furnace speeds. On the contrary, the furnace tonnages have increased. The copper minerals contained in the original ores and concentrates have been mainly chalcopyrite and chalcocite and have assisted in producing a good, fast-smelting sinter at the blast furnaces. The copper sulphides have retarded the lead fusion and permitted a greater degree of zinc sulphide roasting. Thus, the copper sulphides have aided in sintering.

Therefore, high copper on a lead blast furnace, if in the form of a well roasted sinter, will not cause any trouble in the working of the furnace.

Copper in the form of dross, when added to a lead furnace and necessitating a sulphur addition for matte making, will foul a furnace and cause it to work inefficiently. Any sulphides added raw to matte out the copper in combination with zinc will produce a zinc mush, along with accretions and an irregular working of the blast furnace.

Any copper on the furnace charge with the present-day double roasting will go either to the speiss, lead or slag, the amount depending upon the arsenic on charge, as the speiss and the slag losses will depend upon the quality of the sinter.

The blast furnace speiss production will vary from plant to plant, depending upon local conditions and the amounts of copper and arsenic on the charge. I well remember at the El Paso smelter for the years 1939 through 1944, with an average of about 0.6 pct arsenic on the furnace charge, that no speiss was produced with over 1.5 pct copper on charge. However, with under 1.5 pct copper we always showed some speiss.

The copper in the blast furnace speiss can be reclaimed through the dross reverberatory, using the iron arsenide in the speiss as a collector for more copper or else the iron speiss can be ball-milled and roasted for arsenic and the resultant calcine worked up in a copper reverberatory or blast furnace.

The copper slag losses were in no way affected by the speiss-fall.

For our present-day roasting program, the degree of roasting can be ascertained by the amount of copper in the slag. I have a rule of thumb that the copper on charge multiplied by 0.16 should give the percent of copper in the furnace slag. Any amounts below this figure indicate excellent roasting.

The furnace lead with a heavy dross-fall will be hotter and cause extra troubles from the dross separating in transfer pots, along with a greater strain in the lead receiving kettles due to heavy accretions formed on sides of the kettles and necessitating more frequent barring down of the same. Frequently at El Paso, with its heavy dross-fall, I have seen a full 65-ton kettle show about 20 tons of lead after the drossing operations. The reduction in no way will affect the quantity of dross produced.

Blast furnace lead containing its copper in the form of dross and varying from a few percent up to, say, 60 pct will be deposited on the surface of the lead in the drossing kettles and will be worked up in either one of three ways, viz: 1. Fluxing with soda ash in the dross reverberatory. 2. Using the arsenic in the blast furnace speiss as the collector of copper in the dross reverberatory. 3. Fluxing with silica and using a sulphide as a collector for the copper.

The coppery products from the reverberatory, such as the matte and speiss, can be sent direct to a copper department; or the speisses, especially of soda, can be roasted with an irony diluent for arsenic elimination and the resultant calcine sent to the copper department.

The benefits from copper in a lead smelter are as follows:

1. Better sinter and a good furnace tonnage, especially if chalcopyrite.

2. If a baghouse is used to collect the sinter fume, the bag life will be aided by copper as a more alkaline dust will be produced at the sinter machines with slightly more fume.

3. Copper has a great affinity for arsenic and will prevent the arsenic from entering the lead bullion. I have seen the reduction badly off on a lead furnace, even up to 2.5 pct lead in the slag, with 4 pct copper and 2 pct arsenic on the furnace charge, yet the bullion was below 0.1 pct As.

The problems from copper are as follows:

1. Lead wells are apt to be useless and on high copper charges, with the lead removed through the first settlers, the copper will cause a hotter lead and probably result in more frequent tap-hole changes.

2. More dross-fall and creating problems at the drossing kettles.

3. More cost at dross reverberatory to work up the dross.

4. Higher metal losses in dump slags.

Conclusions

1. Copper will not affect the operation of a lead blast furnace, if in the form of a well roasted sinter; but if contained as a dross, definitely it will.

2. Copper will not affect the fume or flue dust production on the blast furnace.

3. Copper will not affect the reduction on a lead blast furnace.

4. Copper will affect the secondaries produced—thus more dross-fall.

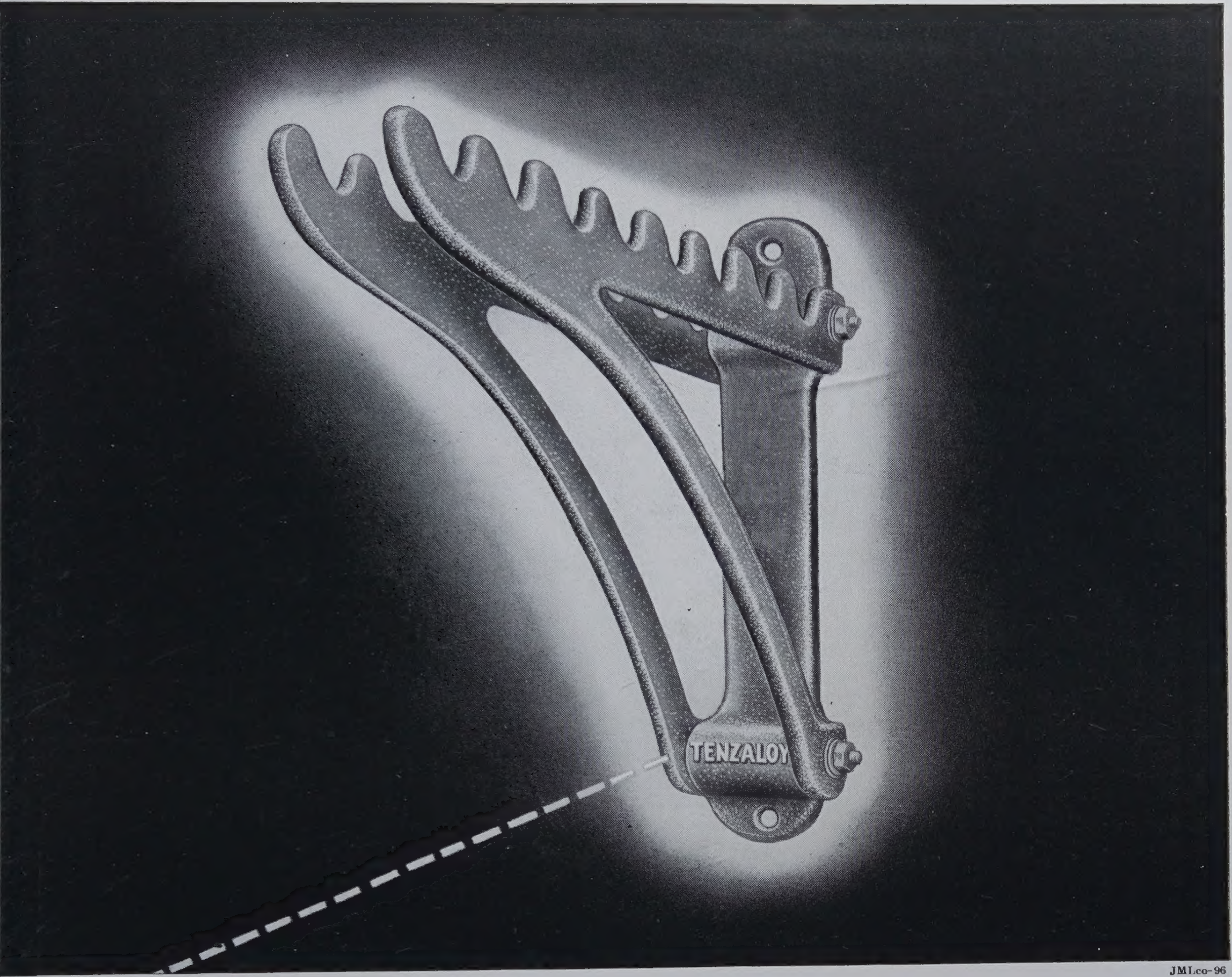
5. Copper will affect the slag losses.

6. Copper will affect and produce a hotter lead.

7. Copper will affect and cause lead-well trouble.

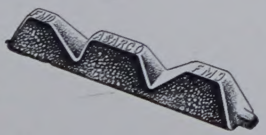
Thus, it is preferable to remove copper at the lead concentrators and make a copper concentrate which can be shipped to a copper smelter. In this way, the copper losses will be considerably lessened, for a lead plant will show possibly a 10 pct copper loss depending upon conditions, against a copper smelter's loss of, say, 2 pct. In addition to the extra losses in the lead plant, one must consider the higher as well as additional processing costs which are incurred when any copper is routed into a lead works, since any copper into a lead smelter must eventually be sent to a copper plant to be worked up into blister copper.

Therefore, to conclude this paper, the writer believes that with the present high metal prices the benefits, such as are obtained in the sinter and the baghouse aids, are apt to be outbalanced by the increased costs for processing and the metal losses.



JMLco-96

For High Strength without Heat Treatment!



This aluminum casting is a tool rack for shovels, hoes, rakes and other heavy garden equipment. The Max Manufacturing Company of San Jose, California, chose Tenzaloy, Federated's new aluminum casting alloy, for the fabrication of the product "because of its great strength as cast, without the necessity of heat treatment." Tenzaloy tensile strengths average 29,000 psi as cast, and 35,000 psi with 10-14 days aging at room temperature.

This manufacturer also states that Tenzaloy is his choice because "standard foundry procedures can be followed in its use, and the end product is a nice clean casting with a minimum of fuss, bother and expense."

Tenzaloy also has excellent ductility, impact resistance, castability, machinability, corrosion resistance and superior polishing characteristics. Send now to Federated's Dept. SJ for literature that describes Tenzaloy more fully.

Federated produces many other non-ferrous metals and alloys, including brass, bronze, aluminum and magnesium ingot; solders; type metals; die casting metals and fabricated lead products. Sales offices in 25 cities across the nation.

Federated first in the non ferrous field FMD

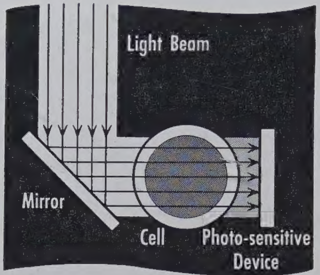
Federated METALS

Division of American Smelting and Refining Company, 120 Broadway, New York 5, N.Y.

NEW The FISHER NEFLUORO-PHOTOMETER

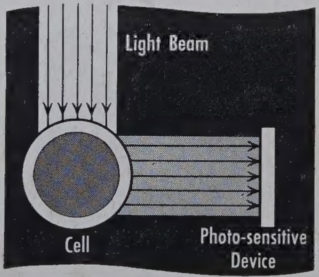
"Three Instruments In One"

1 Colorimeter



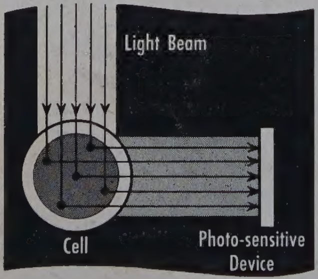
Light of specific wave lengths is absorbed by the unknown in solution. The amount of absorption is measured and related to the concentration.

2 Fluorometer



Light of one wave length causes some substances to fluoresce or emit light (of a different wave length) which is measured at right angles to the incident light and related to the concentration.

3 Nephelometer



Dispersed particles scatter the incident light by reflection and diffraction. This scattered light is measured at right angles and related to the concentration of the particles.

This versatile new instrument permits accurate quantitative determinations to be made by any of the three methods shown at left.

The same few simple controls are used for all three types of determinations and readings for all analyses are made from the combination percentage transmission-logarithmic scale of the instrument.

The Nefluoro-Photometer is quickly interchangeable for colorimetric, fluorometric or nephelometric analysis. Sets of filters, cells, light sources, etc., are available for determinations in the three fields.

Fisher Nefluoro-Photometer (basic instrument only) for 115-volt, 50-60 cycle A.C. **$275.00**

Headquarters for Laboratory Supplies

FISHER SCIENTIFIC CO.
717 Forbes St., **Pittsburgh (19), Pa.**
2109 Locust St., **St. Louis (3), Mo.**

EIMER AND AMEND
Greenwich and Morton Streets
New York (14), New York

In Canada: Fisher Scientific Co., Ltd., 904 St. James Street, Montreal, Quebec